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BRITISH WELDING JOURNAL

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1961 NOVEMBER

VOL. 8 NO. 11

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Inside front cover

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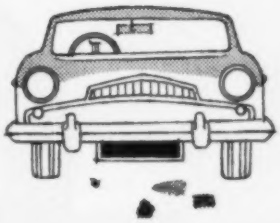

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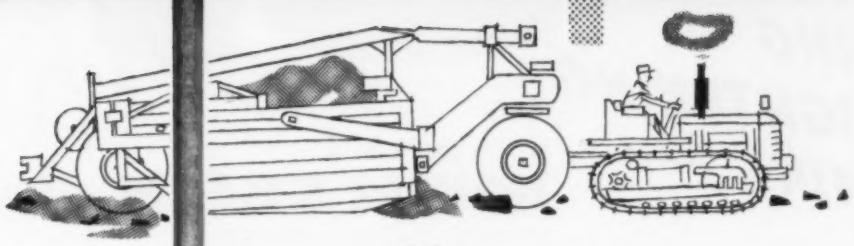

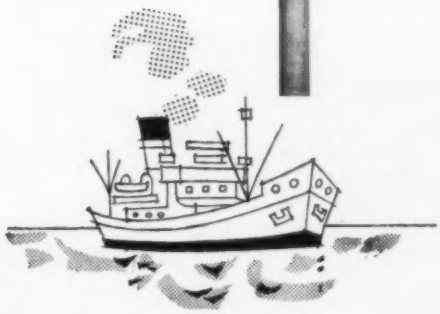

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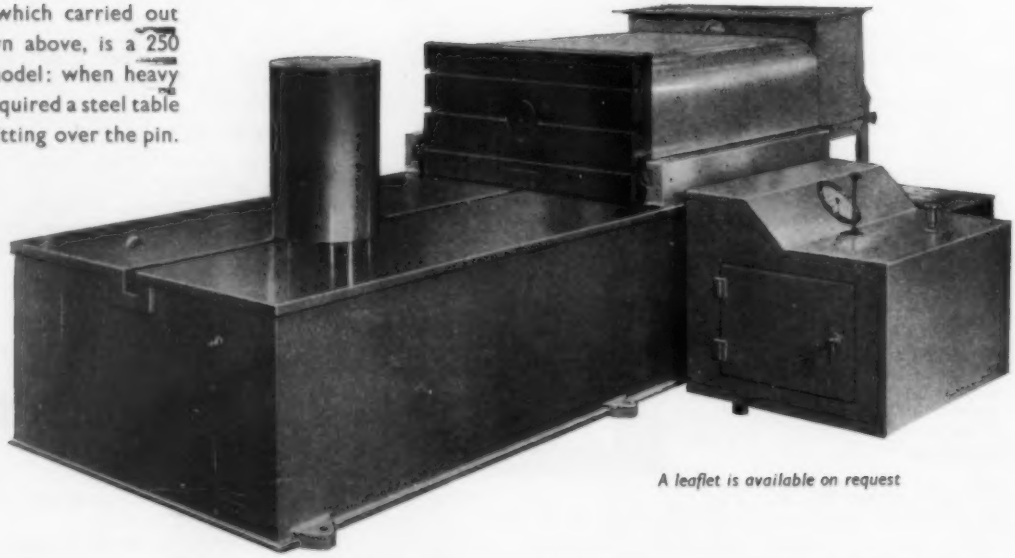

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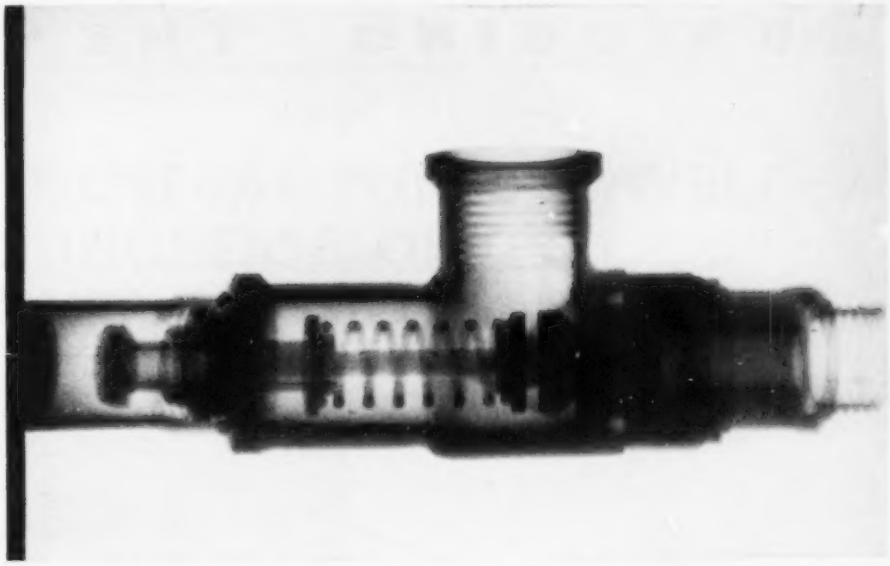

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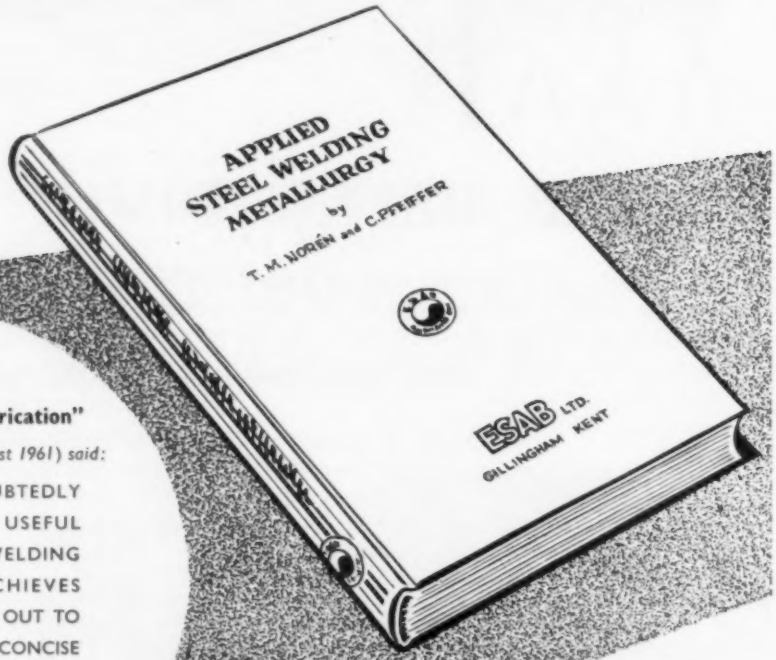

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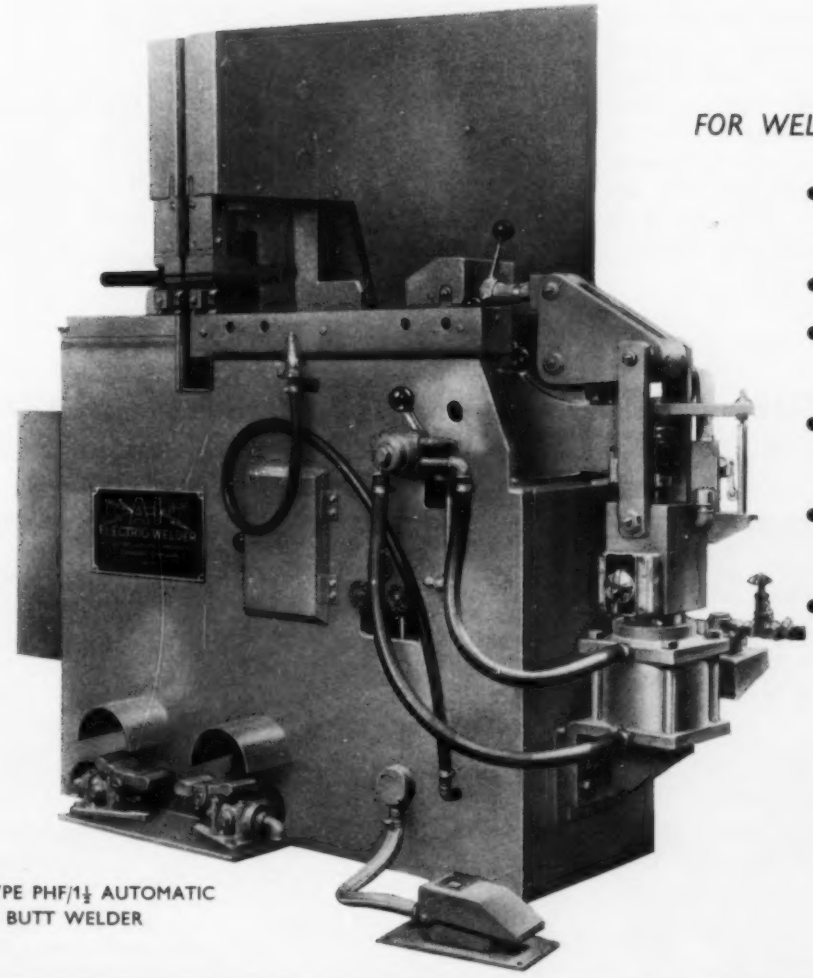

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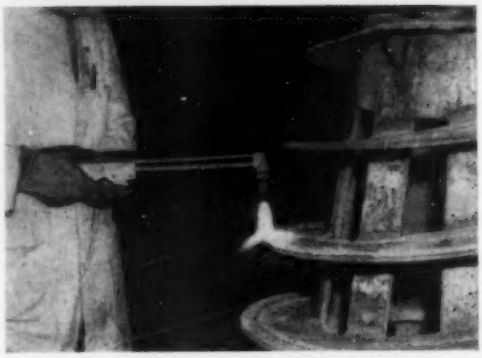

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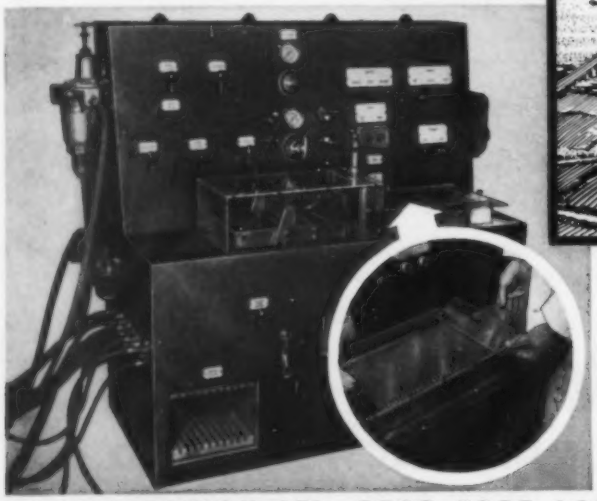

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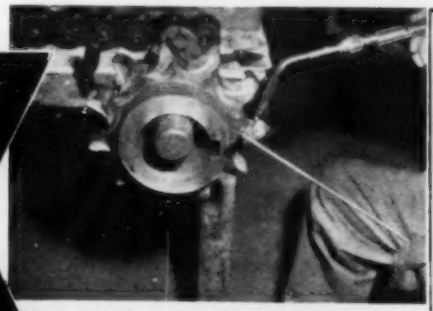

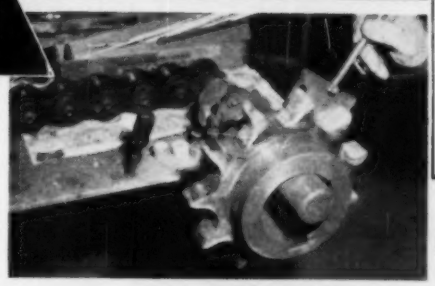

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Above: Tooth wears on driving side

Top left: Application of Colmonoy to steel plate sprocket using oxy-acetylene welding

Bottom left: Checking tooth contour. Reshaping is possible if the profile does not match the template

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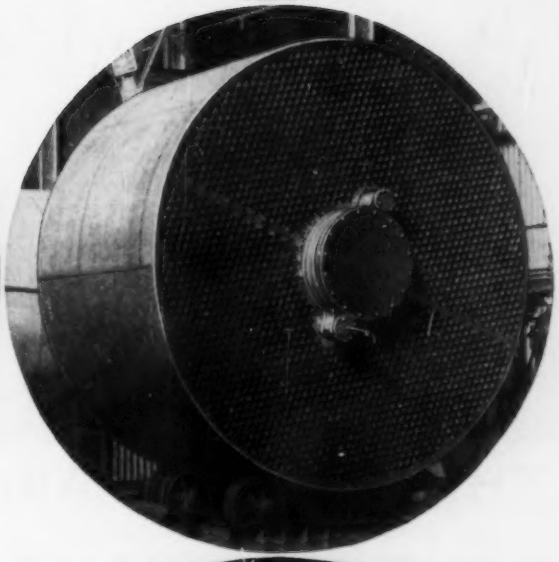

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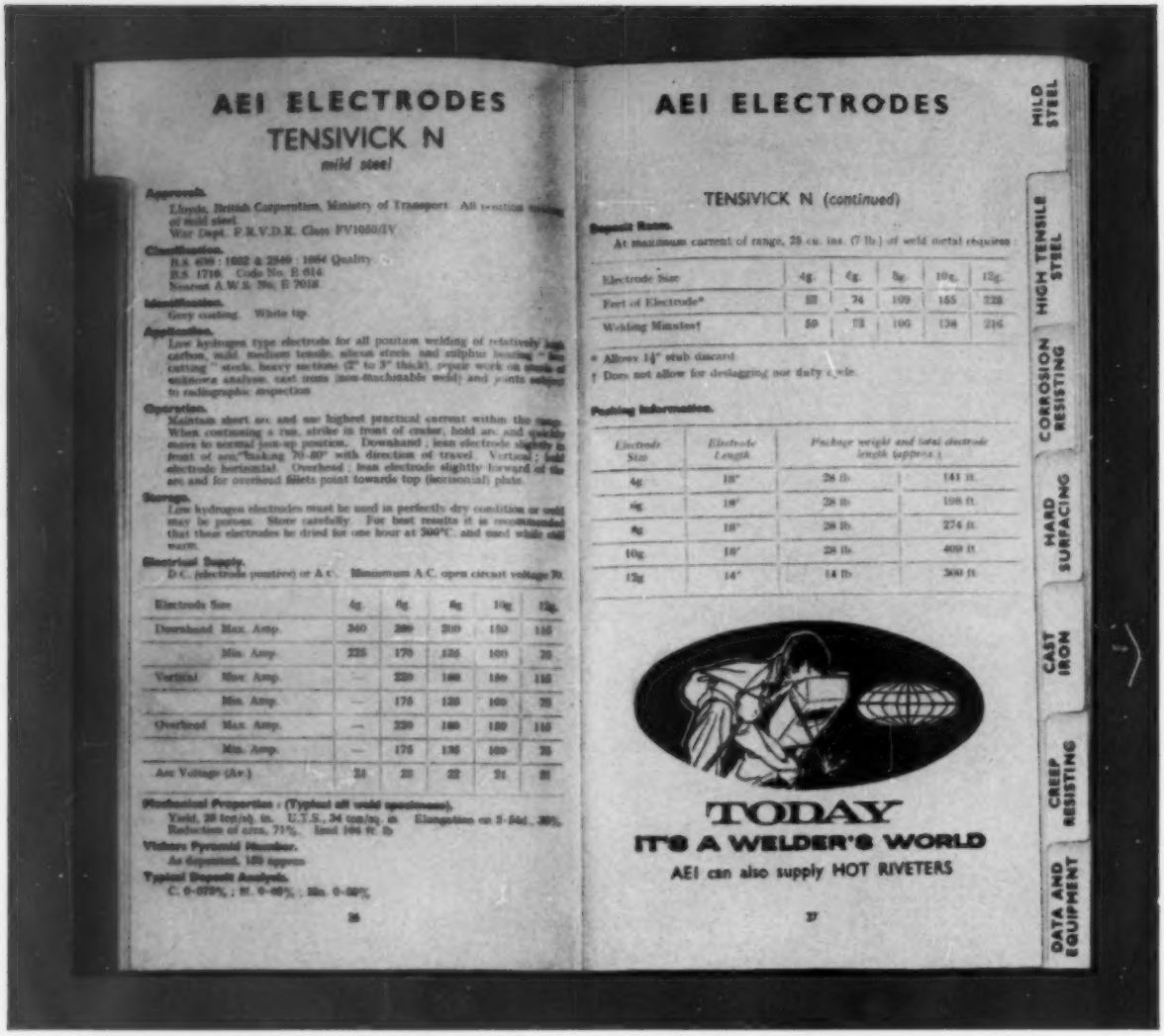
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mild steel
Approvals.
Lloyds, British Corporation, Ministry of Transport. All position welding of mild steel.
War Dept. P.R.V.D.R. Class PV1050/IV.
Classification.
B.S. 639 : 1952 & 2640 : 1954 Quality 3.
B.S. 1719. Code No. E 014.
Nearest A.W.S. No. E 7018.
Identification.
Grey coating. White tip.
Applications.
Low hydrogen type electrode for all position welding of relatively high carbon, mild, medium tensile, silicon steels and sulphur bearing "free cutting" steels, heavy sections (2" to 3" thick), repair work on steels of unknown analysis, cast irons (non-machinable weld) and joints subject to radiographic inspection.
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Maintain short arc and use highest practical current within the range. When continuing a run, strike in front of crater, hold arc and quickly move to normal join-up position. Downhand: lean electrode slightly in front of arc, making 70-80° with direction of travel. Vertical: hold electrode horizontal. Overhead: lean electrode slightly forward of the arc and for overhead fillets point towards top (horizontal) plate.
Storage.
Low hydrogen electrodes must be used in perfectly dry condition or weld may be porous. Store carefully. For best results it is recommended that these electrodes be dried for one hour at 300°C. and used while still warm.
Electrical Supply.
D.C. (electrode positive) or A.C. Minimum A.C. open circuit voltage 70.
Electrode Size 4g 5g 8g 10g 12g
Downhand Max. Amp. 260 280 310 150 145
Min. Amp. 225 170 125 100 70
Vertical Max. Amp. — 220 180 180 145
Min. Amp. — 175 125 100 70
Overhead Max. Amp. — 220 180 180 145
Min. Amp. — 175 125 100 70
Arc Voltage (Av.) 24 22 22 21 21
Mechanical Properties: (Typical all weld specimens).
Yield, 28 ton/sq. in. U.T.S., 34 ton/sq. in. Elongation on 5·64d, 30%.
Reduction of area, 71%. Izod 106 ft. lb.
Vickers Pyramid Number.
As deposited, 189 approx.
Typical Deposit Analysis.
C. 0·07%; Si. 0·40%; Mn. 0·80%.
35
AEI ELECTRODES
TENSIVICK N (continued)
Deposit Rates.
At maximum current of range, 25 cu. ins. (7 lb.) of weld metal required:
Electrode Size 4g 5g 8g 10g 12g
Feet of Electrode* 82 74 109 165 228
Welding Minutes† 59 73 106 138 216
* Allows 1½" stub discard.
† Does not allow for deslagging nor duty cycle.
Packing Information.
Electrode Size | Electrode Length | Package weight and total electrode length (approx.)
4g 18" 28 lb. 141 ft.
5g 18" 28 lb. 198 ft.
8g 18" 28 lb. 274 ft.
10g 18" 28 lb. 409 ft.
12g 14" 14 lb. 300 ft.
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THE RESULT—HIGH SPEED HIGH QUALITY SUBMERGED ARC WELDS

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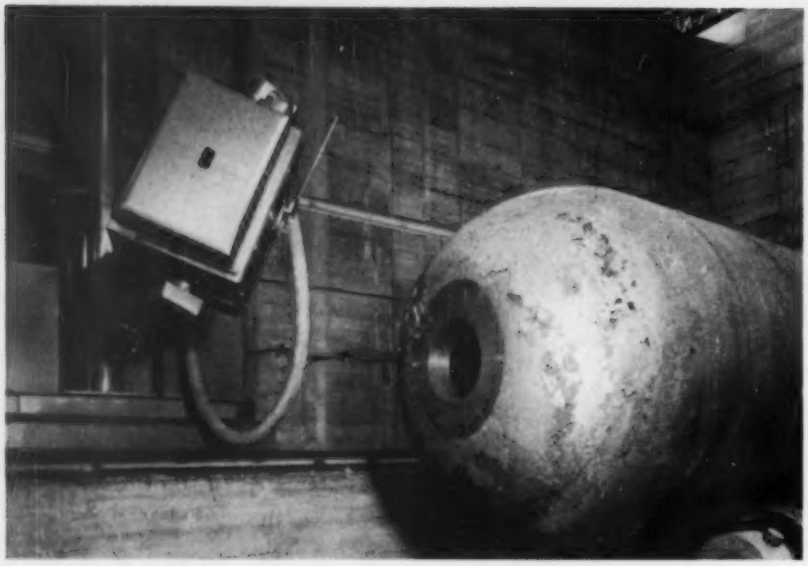

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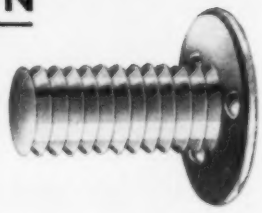

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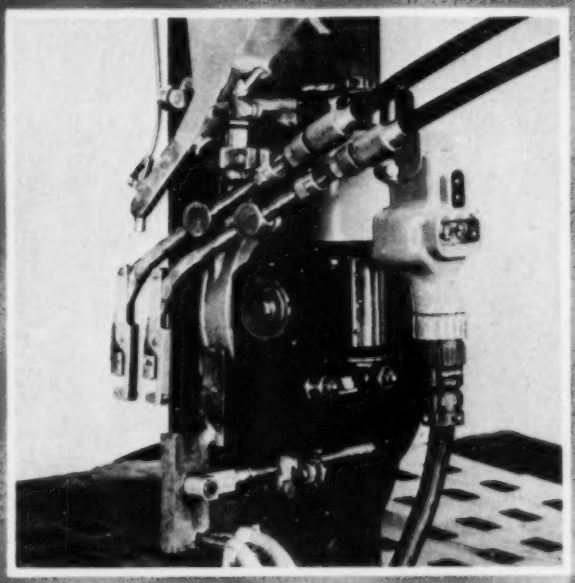
Automatically - ESAB



The **ES 5** machine shown alongside is

The **ONLY** available

PORTABLE electro-slag

WELDER

The welding head weighs less than 100 lbs.— and can therefore be taken to the work.

Constructed to the specification as devised and perfected by Prof. Boris Paton of the Russian Institute bearing his name, the ESAB ES5 is capable of welding steel plates ¾" to 3⅛" thick, vertically, in 1 pass. Each machine is equipped to make vertical fillet and butt joints.

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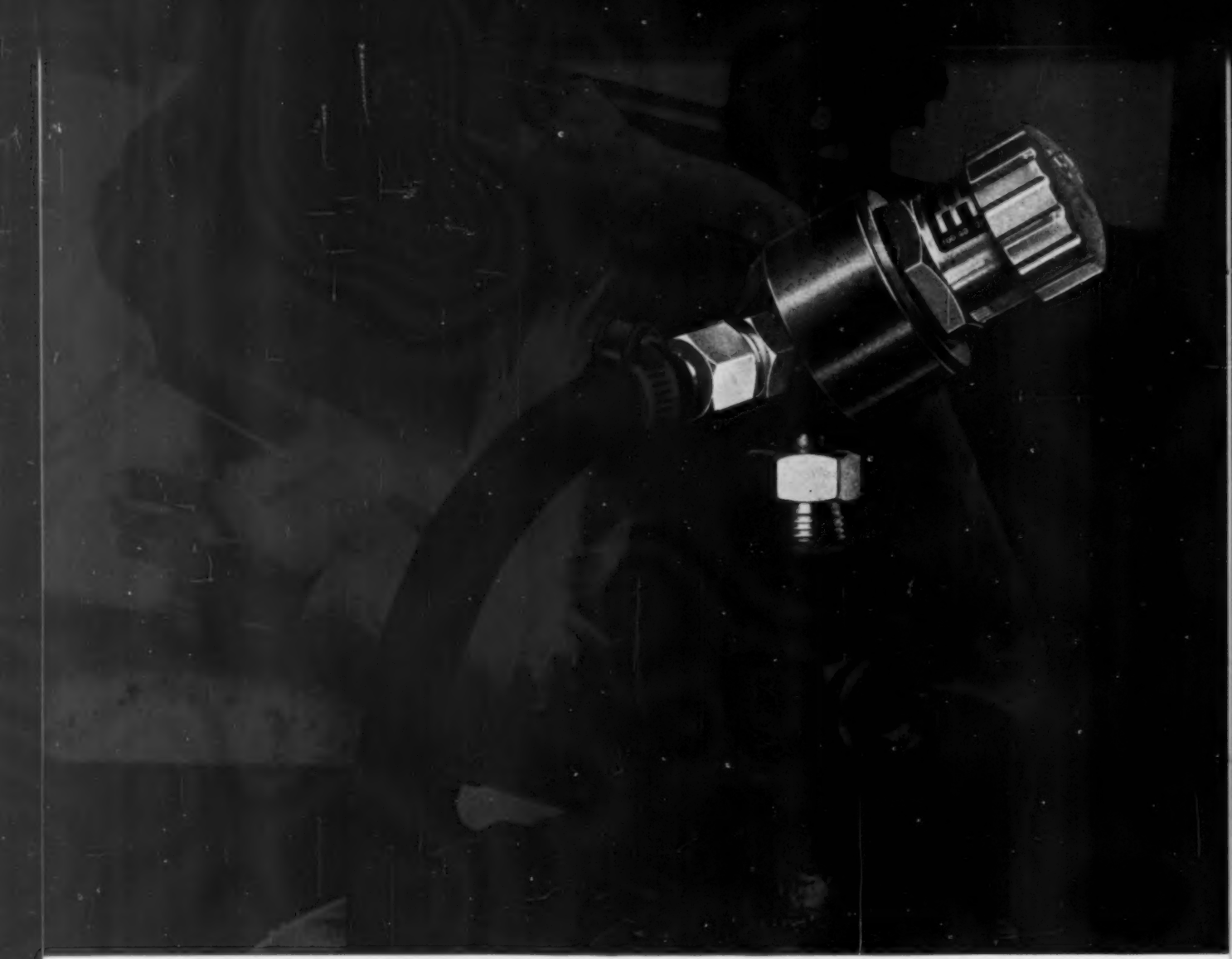

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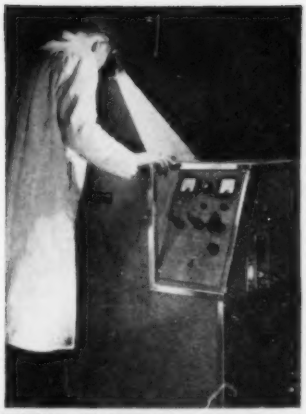

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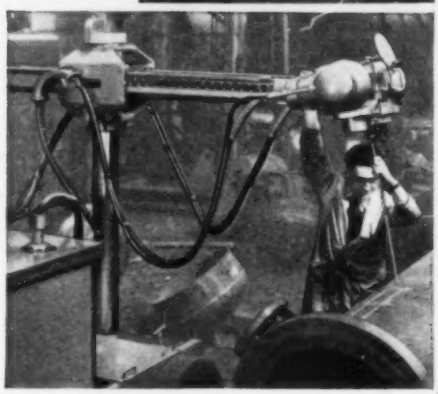

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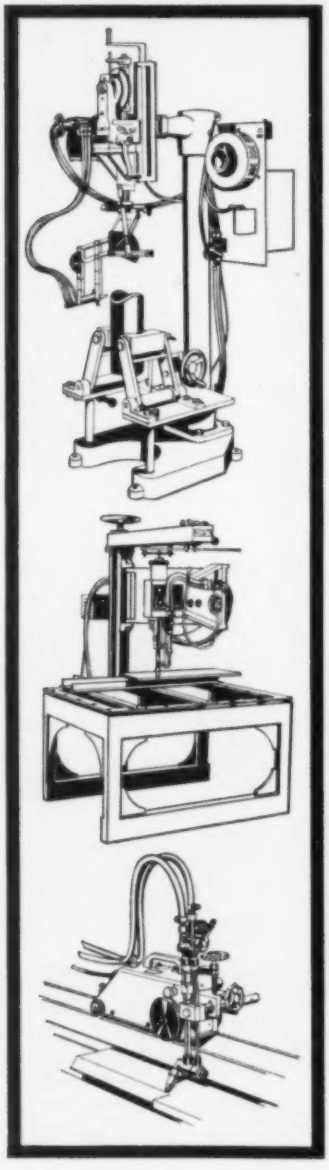

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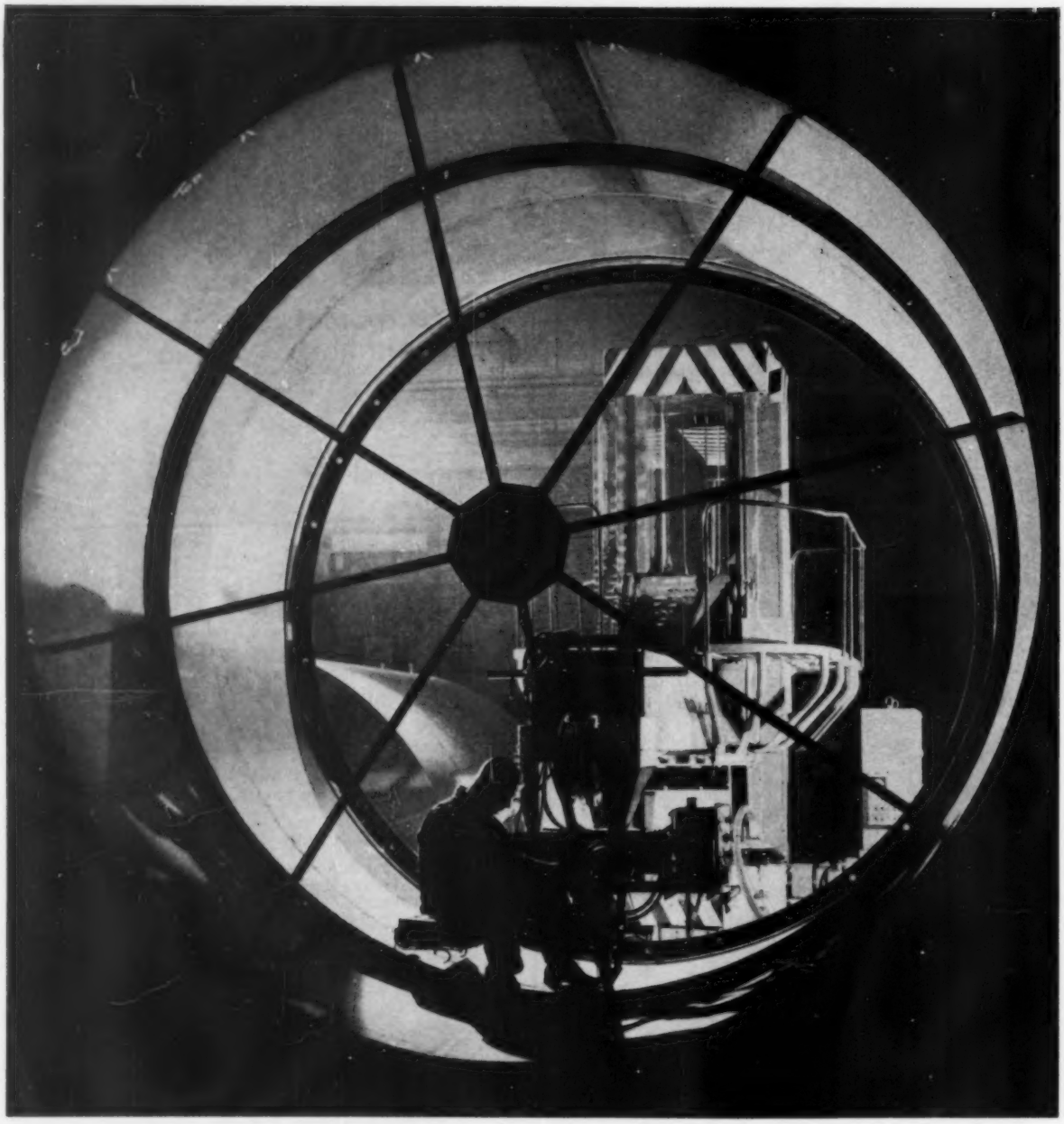

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Welding TECHNOLOGY

A Note on the Welding Behaviour of Niobium and Vanadium

By R. D. Johnston, B.SC., A.I.M., A.M.INST.W. *and L. W. Derry*, B.SC., M.SC., F.I.M.

The authors give a preliminary explanation for a phenomenon experienced in the welding of Nb sheet to V sheet. The weld-parent metal interfaces were in the form of long 'tongues', and the shape was affected by the welding current and traverse speeds. The effect is thought to be due to the dissolution of Nb in molten V, but further work is being carried out, also on other systems, to determine the influence of alloying properties as indicated by phase equilibrium diagrams.

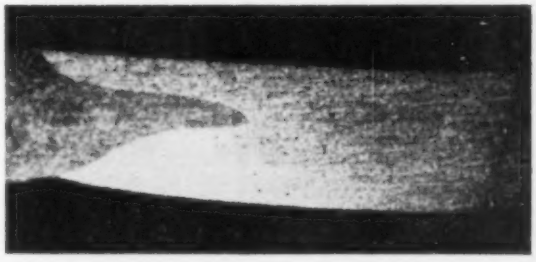
DURING some recent investigations into the welding of niobium to vanadium, an unusual effect was noticed at the interface between the niobium parent plate and weld metal.

The specimens were welded with a straightforward fusion butt weld, no filler metal being used, the source of heat being a traversing tungsten electrode and d.c. arc. An argon atmosphere covered the weld entirely at all times.

The specimens were made from $\frac{1}{16}$ in thick sheet, $\frac{3}{4}$ in. wide and 3 in. long. The abutting edges were prepared simply by filing and grinding until they were square, and were not shaped in any way. They were placed with the prepared edges touching and overhanging a central channel in a copper backing plate. Spring clips held the specimens in position.

On sectioning the welds, it was noticed that, whilst the vanadium–weld interface showed an apparently normal contour, the niobium–weld interface exhibited an entirely different shape. This is clearly seen in Fig. 1.

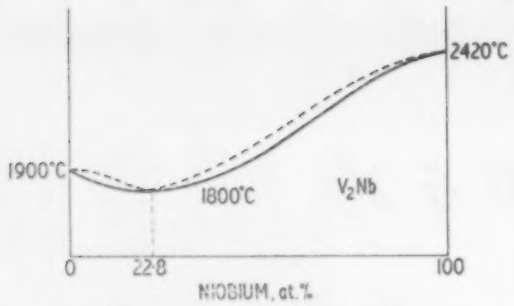
No similar effect has been found for other metals, as far as it is possible to establish following examination of available literature, and it is for this reason that the effect is reported here. It will be referred to subsequently as the 'tongue effect'. However, apex-shaped interfaces were observed by Theilacker, Baugh, Kasburg, and Stermon[1] when welding hafnium to zirconium, though the form was rather different and no satisfactory explanation was given.



1—*Appearance of 'tongue effect' at Nb–V weld junction*

The tongue effect in niobium-vanadium welds was first noticed when specimens were sectioned following experiments designed to relate beadwidth to current and traverse rate. A series of specimens was welded at a constant current of 70 amp and at traverse rates varying from $\frac{1}{2}$ to 4 in./min in $\frac{1}{2}$ in./min stages.

Manuscript received 15th December 1960.
The authors are at the Battersea College of Technology, London.

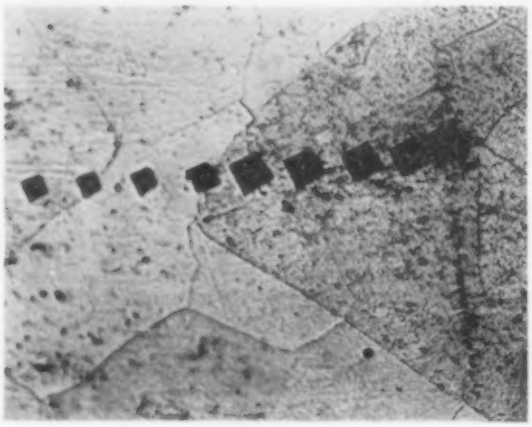


2—*Equilibrium diagram for the Nb–V system*

Porosity appears to occur particularly at the boundary between the vanadium and weld metal, and is attributed to the high oxygen content (1800 p.p.m.) of the vanadium. The niobium is low in oxygen and thus no porosity is found in association with the niobium tongue.

The material of the molten zone is an alloy of as yet unknown composition. Experiments are in hand to determine this. It is considered, however, that the composition will be fairly near to that of the lowest melting alloy of the system Nb/V, which forms a continuous series of solid solutions with a minimum at 22·8 at.% Nb and 1810°C. The equilibrium diagram[2] is shown in Fig. 2.

It is thought that the temperature of the weld pool is not far above the melting point of vanadium, and that the molten material is in a state of constant agitation, which ensures an even temperature and homogeneous composition. Subsequent micro-hardness tests support this view. Figure 3 shows a hardness traverse carried out through the tip of the tongue and indicates that there is no gradation in hardness as the boundary is approached on either side. The change in hardness is sudden, and then a constant level persists on both sides of the interface. This would seem to indicate that there is no diffusion layer showing a gradual fall in vanadium content in the solid niobium, but an extremely sharp division between niobium and weld metal. The same conditions seem to be true at the vanadium end of the weld. It will be noticed also in reference to the niobium that as the weld solidified, the new grains grew onto the remains of the existing ones in the niobium.



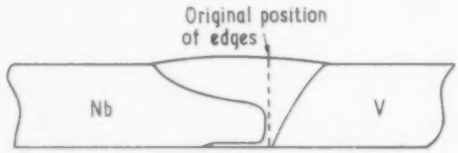
3—*Hardness traverse of tip of tongue (30 g load)* ×300

Microhardness tests carried out across the tip and parallel to the interfaces gave similar results to those already mentioned.

Generally, it would be assumed that the boundaries between weld metal and parent plate represent temperature contours, where the temperature would be that of the melting point of the parent plate. However, in this case, it is thought that, whilst this is probably true at the vanadium interface, it does not apply at the niobium interface. Here the contours are governed by the dissolution of niobium in vanadium, remembering that the weld pool temperature may be below the melting point of niobium. Thus, on the niobium side there is a solution boundary or front, and not a temperature boundary.

The sectioning of further specimens has shown that at higher electrode traverse rates for a given current, or similarly at lower currents and a given traverse rate, the tongue appears to have a much more blunt outline (*see* Fig. 4).

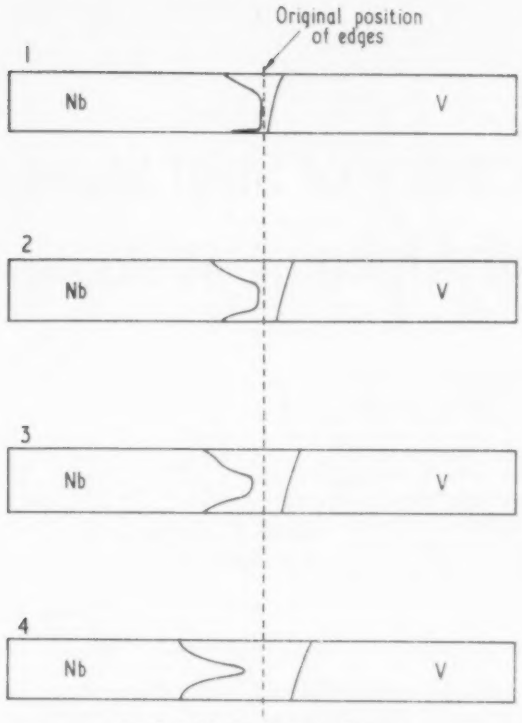
Sectioning a series of specimens showed that there was a definite trend from the blunt form to the sharp form of Fig.1. In addition, it was noticed that at the start of a weld it was necessary to melt the vanadium first before the niobium would melt; in fact, a true bond would not form until the vanadium had run into the niobium and formed an alloy. Once this had happened welding appeared to continue normally. This helps to substantiate the contention that the temperature of the weld metal is below the melting point of niobium, but above that of vanadium, though the exact nature of the relationship to the vanadium melting point is not yet known accurately. (It was found that to weld niobium to niobium, 30% higher currents were required than to weld vanadium to vanadium, for a similar bead width. This would be expected, taking into account the melting points and thermal conductivities of the two metals.)



4—*Blunt outline of tongue at higher traverse speeds*

Thus the evidence suggests that the blunt tongue is an early stage in the development of the more sharp outline. The sharp contour is reached fairly quickly in the series and then is a constant feature; *e.g.*, a current of 70 amp was used in the series of welds where the traverse rate varied from ½ in./min to 4 in./min. At a speed of 3 in./min, the tongue became blunt, and at 4 in./min the tongue had the shape indicated in Fig. 4.

It is unlikely that the explanation of the tongue effect lies in conductivity or specific heat considerations, though melting temperature differences are important. The probable mechanism by which the



5—*Probable development of tongue*

tongue is formed is one of dissolution of niobium in molten vanadium. The vanadium, melting first, quickly forms an alloy by dissolving a portion of the top surface of the niobium. This weld metal, in a high state of agitation and probably high fluidity, either falls or is forced down between the plate edges, and wets the underside of the niobium. The vanadium is melted back simultaneously and the condition shown in Fig. 4 would be reached. Further dissolution of niobium occurs, probably at a rate at which the alloy composition is kept constant, though faster at corners than at faces. Thus the dissolution contours would probably be similar to those shown in Fig. 5, resulting in a tongue being formed.

It is probably true that vanadium is also dissolved by the weld metal at a later stage, but since it is likely that the weld temperature is above the melting point of vanadium by about 200°C., it would be more of a melting effect, whereas the niobium is definitely being dissolved if the hypothesis is correct.

It would seem that the suggested mechanism provides a likely explanation of the shape of the interface. Work is continuing with the aim of establishing the mechanism, and explaining it in terms of kinetics; in addition, it is intended to experiment with other systems so as to establish whether alloying properties as indicated by the phase equilibrium diagrams have any specific effect on tongue formation. It is intended also to mask off the underside of the niobium and so prevent it being wetted by the weld metal, with subsequent dissolution from the underside.

Acknowledgments

The authors would like to thank the Fellowships Committee of The British Oxygen Company and the U.K.A.E.A. for permission to publish this note. Special thanks are due to Mr. J. B. Chater and his staff at the Research Division of B.O.C., and to Mr. A. F. Taylor of the U.K.A.E.A. for generous advice and co-operation.

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2. — Hansen: "Constitution of Binary Alloys", ———, 1958.

BOOK REVIEW

J. A. Oates: "*Modern Arc Welding Practice*", London, 1961, George Newnes Ltd., 293 pp (30*s*.)

Welding is now firmly established as one of the cornerstones of modern technology. Newcomers to the field are sometimes rather overwhelmed by the scope and diversity of the specialised knowledge that has to be absorbed. "*Modern Arc Welding Practice*" is stated by the author to be a practical reference and instruction manual and will be extremely useful to those entering the welding industry at any level as well as being a useful text book for the student welder.

Chapters are included on most of the aspects of arc welding practice such as welding plant, electrodes, flame and arc cutting, work preparation, etc. The welding of materials other than mild steel receives detailed attention. Separate chapters are devoted to the welding of cast irons, aluminium alloys, copper and nickel alloys, stainless steel and hard facing techniques. Argon arc welding is also fully dealt with in a comprehensive manner.

The author has confined himself to a descriptive treatment of the subject matter, the theoretical and metallurgical aspects of welding being kept to the minimum necessary for clarity. It is very easy for a book of this type to fall between two stools and the author is to be congratulated on providing a straight forward descriptive handbook. The style is direct, easily readable and the subject matter contains a great deal of common sense. When the welding of armour plate is mentioned, for example, the reader is reminded that the electrode must comply with the relevant ministry specifications, a point which, from the practical aspect, is quite as important as a background knowledge of the theory of hydrogen cracking. The subject matter is amplified by clear line drawings where necessary and many excellent illustrations of specialised equipment are included. The readers attention is also drawn to the information contained in the appropriate British Standard.

Whilst the treatment of the subject matter is simple and concise and the book is by no means a theoretical text book on welding it can be thoroughly recommended to the graduate entering the welding field for the first time as well as to the student welder.

J. MERCER

Determination of the Hot Crack Susceptibility of Steel Sheet during Argonarc Welding

A simple test for determining the comparative hot-crack susceptibility of steel sheet is described and a method of deriving a comparative scale of potential crack susceptibility from the test is suggested.

Results of tests on four alloy steels are given, two of these results being of considerable significance; an air-melted 2% Ni–Cr–Mo steel which showed marked susceptibility to cracking and a high purity vacuum melted steel of the same specific composition which showed a potential crack susceptibility of zero. It is suggested that a study of the complete compositions of these two steels would provide valuable information on the constituents that encourage susceptibility to hot cracking.

By H. V. Huxley,
A.M. INST.W.

Introduction

The measure of a steel's weldability might well be defined as the degree of susceptibility it shows to 'hot' cracking or tearing in the region of the weld, for, of the factors which determine weldability, this is undoubtedly the most important. Hot cracking, which almost certainly occurs at a temperature in the region of the solidus, probably reflects inadequate ductility at that temperature and therefore inability to accommodate the thermal strains set up during welding.

Many forms of test have been devised to determine resistance to hot cracking, examples being the Double Fillet Test[1] and the Houldcroft Test.[2] It is apparent that the development of a simple yet sensitive test would greatly facilitate study of the factors which cause or contribute to hot cracking.

The value of any test which relies on measurement of either the number of cracks formed or on crack length, when a weld or melt run is made on a selected steel under controlled conditions, can best be assessed by comparing the results with the extent of cracking experienced during production welding of the same steel. The establishment of such a relationship generally presents difficulty, for, no matter how carefully a production welding process may be controlled, some variables are bound to remain and therefore the relationship is unlikely to be precise.

A weldable low-alloy steel, which has been used extensively in sheet form for the fabrication of rocket motor cases and pressure vessels, and which is capable of achieving strength levels in excess of 80 tons/sq.in. by a suitable heat treatment, is the 1% Cr–Mo steel to Bristol Aerojet Limited specification RS.120. Experience in the welding of this steel by the argon-shielded tungsten-arc process shows it to vary widely between batches in susceptibility to hot cracking, the marked susceptibility found in some batches being attributed to relatively high sulphur and phosphorus contents.[3]

The existence of a number of batches of this steel sheet, each with a well documented history of weldability in production, has greatly facilitated study of the phenomenon of hot cracking. One result of this study has been the development of a test for determining the hot cracking resistance of steel sheet during welding by the argonarc process, excellent correlation being obtained between weldability as indicated by the test and that shown by experience in production.

The test has been used to determine the crack susceptibility of four alloy steels, two of them giving results of particular interest.

Description of Test

Three forms of the test specimen are illustrated in Fig.1. This specimen takes the form of a strip 16 in. long by 1½ in. wide having opposing slots approximately $\frac{1}{32}$ in. wide cut into the sides to a depth of 0·6 in. These slots serve a double purpose, acting as thermal insulators between each test section and giving an edge starting effect for the initiation of a crack. In the experimental forms the distance between each pair of slots was 1 in. (specimen *A*), 1½ in. (specimen *B*) and 2 in. (specimen *C*).

To carry out a test the specimen is set up on a special jig (Fig. 2) consisting essentially of two parallel

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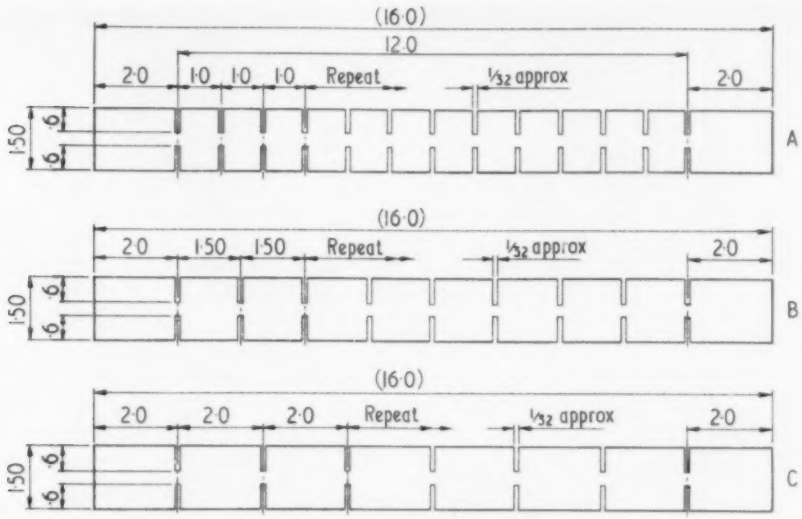
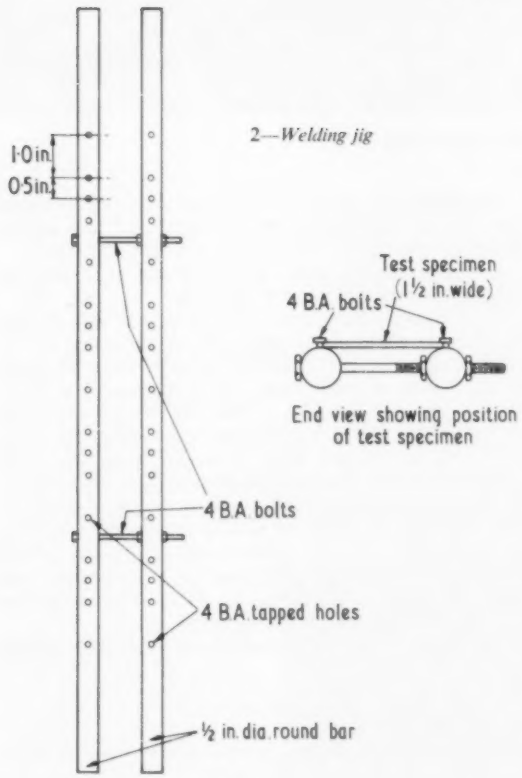


1—*Crack test specimens*



2—*Welding jig*

round steel bars ½ in. dia. by about 18 in. long. The specimen is laid on the jig in such a way that line contact only is maintained between specimen and jig; this ensures a minimum loss of heat to the jig by conduction. The specimen is held loosely at the edges by screws let into the jig at suitable intervals, the points of retention being at the ends of each test section. This arrangement allows movement, due to thermal stresses, to be unrestrained. A melt run is made along the length and in the middle of the specimen, care being taken to avoid 'bridging' the ends of the slots.

Figure 3 shows the assembly set up on a welding machine on completion of a test.

The mechanism of the test would seem to be as follows: The area of metal between each pair of slots is, as nearly as possible, thermally insulated from its neighbours and from the jig. As the arc enters this area and proceeds towards the next pair of slots the cooling melt behind the arc is subjected to strain due to its own contraction against the restraint of the solid and comparatively cold metal ahead of the arc. With the forward movement of the arc and the gradual build-up of heat within the test area the restraint, and therefore the strain on the cooling melt run, decreases progressively. At the starting slots the strain is sufficient to initiate a crack. The extent of the propagation of this crack (crack length) under decreasing strain, is the measure of the steel's hot-crack susceptibility.

Evaluation of Test

Five batches of the RS.120 1% Cr–Mo steel sheet of a nominal 15 s.w.g., which had shown a wide variation in susceptibility to hot cracking during production welding, were examined by means of the test. The compositions of these steels, together with the requirements of the specification, are given in Table I.

It was assumed that to obtain maximum sensitivity the test section length (distance between slots) would need to be as large as possible, and in the early stages of development tests were carried out on specimens having distances up to 4 in., welding speed being constant at 5 in./min. As was to be expected, the overall scatter in crack length became wider as the test section length increased. It follows that with a

3—*Assembly at completion of welding*

Table I
Composition of 1% Cr–Mo steels

Steel No.	*Composition, %*							
	C	*Mn*	*Si*	*S*	*P*	*Ni*	*Cr*	*Mo*
1	0·30	0·49	0·29	0·009	0·016	0·18	0·96	0·19
2	0·295	0·44	0·24	0·011	0·015	0·13	0·90	0·24
3	0·315	0·41	0·19	0·015	0·030	0·09	0·85	0·245
4	0·29	0·51	0·14	0·016	0·023	0·16	0·95	0·195
5	0·315	0·41	0·19	0·013	0·031	0·09	0·82	0·245
Requirements of RS.120								
Min.	0·30	0·40	0·10	—	—	—	0·80	0·15
Max.	0·35	0·60	0·35	0·015	0·025	0·30	1·10	0·25

wider scatter in crack length the greater is the number of tests needed to establish a value of crack sensitivity commensurate with maximum economy in the use of material. The ultimate choice of test section length was therefore restricted to one of the dimensions shown in Fig. 1.

To obtain 24 test sections for each form of specimen (*see* Fig. 1) two specimens were prepared to form *A*, three to form *B* and four to form *C* in each steel.

In each case the melt run was made at a speed of 5 in./min using a current of 80 amp.

The results of the tests are given in Table II, the relative hot-crack susceptibility being indicated by the average crack length obtained on each form of specimen. Table II also shows the production welding history of each steel, *i.e.*, the percentage of welds found to be cracked when examined radiographically.

It will be noticed that in many cases the total number of cracks measured was less than the nominal 24. This is because the melt run had sometimes diverged and over-run the end of a slot, so invalidating the following test section. Where such divergence of the melt run takes place, fracture usually occurs and extends from the slot into the melt, so altering the pattern of stress within the test section.

The crack sensitivity of the five steels, as determined

Table II
Crack tests on 1% Cr–Mo steels

Steel No.	*Crack Test Results*						*Production Weldability*	
	Specimen A		*Specimen B*		*Specimen C*			
	No. of Cracks	*Average Crack Length, in.*	*No. of Cracks*	*Average Crack Length, in.*	*No. of Cracks*	*Average Crack Length, in.*	*No of Sheets Welded*	*Welds Cracked, %*
1	22	0·201	24	0·280	21	0·299	364	4·2
2	22	0·203	24	0·321	23	0·344	59	11·8
3	23	0·383	24	0·554	23	0·637	84	32·2
4	21	0·416	24	0·560	24	0·803	188	43·5
5	18	0·503	22	0·615	22	0·943	87	64·5

by the three forms of test and as indicated by production experience, is compared graphically in Fig. 4.

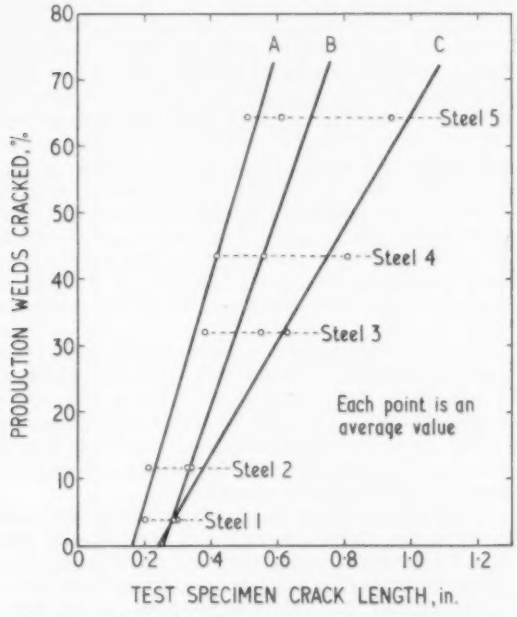
Figure 5 shows two specimens to form *C* of Fig. 1 after welding. The specimens were prepared from two batches of the 1% Cr–Mo steel (Table I) which had widely differing histories of weldability in production (Nos. 1 and 5 of Table II). The difference in crack length is clearly visible.

A study of the test results on the five steels of Table I suggested that, for a given steel, the average crack length obtained with each form of specimen was very nearly the same proportion of the distance between slots in each case. In Fig. 6 the production welding history of the five 1% Cr–Mo steels (Tables I and II) is plotted against the average crack lengths obtained on each of the three forms of specimen, expressed as a proportion of the distance between slots. Whilst some scatter is evident this is, perhaps, no greater than might be expected in a test of this type.

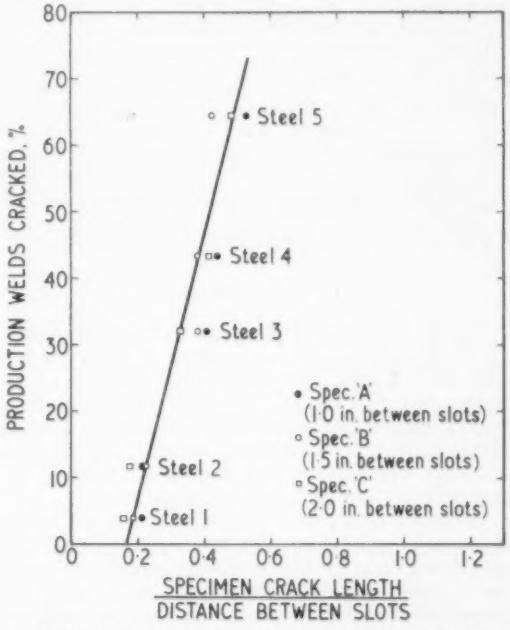
Tests on Alloy Steels

The steels tested were a 3% Cr–Mo–V, a 5% Cr–Mo–V, and two Ni–Cr–Mo steels to specification SAE.4335. Of the two latter steels, one was a normal air-melted cast and one was manufactured by vacuum melting using a carbonyl-iron base and pure metal additions with no pick-up of impurities during melting. All steels were in sheet form and of a nominal 15 s.w.g. The compositions of the steels are given in Table III.

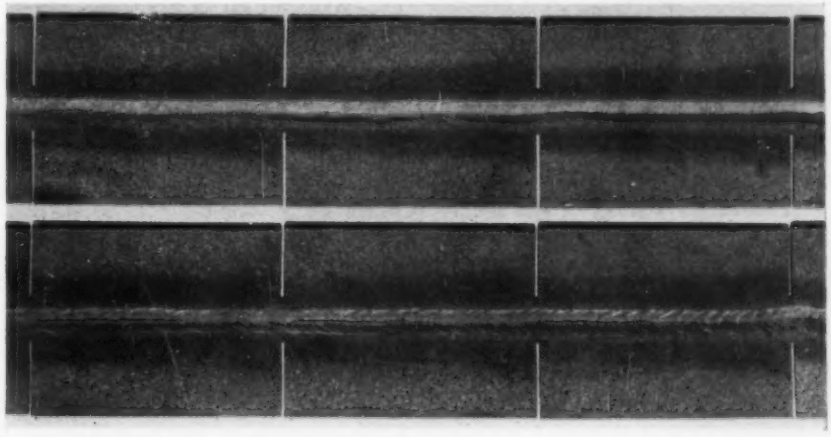
Test specimens were prepared from each steel to the dimensions shown in Fig. 1 (*B*) (1½ in. between slots), it



4—Comparison of crack sensitivity of 1% Cr–Mo steels



6—Relationship between production experience and crack test results



5—Specimens C after welding

Table III
Composition of alloy steels

Steel No.	Specification	Composition, % C	Mn	Si	S	P	Ni	Cr	Mo	V	Cu	Sn	As
6	DTD.730 (Analysis)	0·38	0·72	0·24	0·005	0·014	0·14	3·08	0·97	0·20	—	—	—
7	MOG.510	0·37	0·41	1·11	0·011	0·016	0·21	4·84	1·41	0·56	—	—	—
8	SAE.4335 (Air melted)	0·37	0·62	0·14	0·012	0·016	1·82	0·82	0·23	—	0·05	0·005	0·021
9	SAE.4335 (Vacuum melted)	0·38	0·59	0·21	0·004	0·001	1·96	0·77	0·27	—	Trace	0·001	Trace

being considered that this dimension offered the best compromise between sensitivity and economy in material. Melt runs were made in the specimens at a speed of 5 in./min using a current of 80 amp.

Table IV gives the results obtained on the four steels.

Table IV

Crack tests on alloy steels

Steel No.	Specification	Number of Test Sections	Average Crack Length in.
6	DTD.730 (Analysis)	15	0·339
7	MOG.510	11	0·278
8	SAE.4335 (Air melted)	16	0·594
9	SAE.4335 (High purity VM)	8	0

Discussion

From the aspect of practical welding, 'weldability' is incapable of precise definition, the observed 'weldability' of a given steel being affected by technique and probably by the type of fabrication involved. However, a steel may be said to possess a certain potential 'weldability' or susceptibility to hot cracking which may be derived from a study of its composition and/or metallurgical factors.

A theory, now widely accepted, is that hot cracking is caused mainly by the presence of quite small quantities of constituents whose melting points are lower than the equilibrium solidus of the steel. These constituents segregate to the grain boundaries and, being liquid, form regions of poor cohesion between grains, the resulting loss of 'ductility' occurring at a time when the strains due to thermal gradients and contractions are high. If it is accepted that the problem is one of adequate ductility over a fairly limited range of temperatures then, to determine the potential hot-crack sensitivity of a steel, some test or device is required which will give a relative measure of the ductility within this temperature range. By reason of its design and mechanism of crack initiation and propagation the crack test would seem to provide such a device.

Production welding of the five 1% Cr–Mo steels was carried out under carefully controlled conditions. The type of fabrication involved was the same in every case and welding speed was maintained at 5 in./min. It is evident from Figs. 4 and 6 that the results given by the test are a very good reflection of production experience in the welding of these steels.

Figure 6 shows that the average crack length developed in a given steel tends to be a constant proportion of the test section length or distance between slots. This suggests that the value of strain developed and the pattern of the action of these strains on the cooling melt are the same in each form of specimen. Reference has already been made to the effect of increasing test section length, it being noted that as this dimension increases so the scatter in crack length becomes wider. This could be due to the greater area for dispersal of heat given by the longer test section length, and it may be that the scatter could be reduced by increasing the welding speed, so giving a faster input of energy to compensate for the faster dispersal of heat by radiation and/or conduction. Such a supposition would imply that an optimum welding speed exists for a given test section length at which the scatter in crack length would be at a minimum. Whilst Fig. 6 would suggest that the average crack length developed in a given steel tends to be a constant proportion of the test section length, the variation in this dimension (1–2 in.) may not be wide enough to display the full effect of varying test section length and, therefore, of overall scatter in crack length.

The test is designed to give a purely comparative measure of a steel's susceptibility to hot cracking and it has been demonstrated that using test section lengths of 1 in., 1½ in. and 2 in., with a welding speed of 5 in./min, the indicated crack susceptibility given by the test correlates very well with experience gained in practical welding at the same speed.

Since there would seem to be a constant relationship between test section length and the average crack length obtained it is suggested that, when the specimen is of a nominal 15 s.w.g., the test section length is not greater than 2 in., and welding speed is 5 in./min, a scale of susceptibility to hot cracking might be derived from the expression:

$$\frac{\text{Average Crack Length}}{\text{Test Section Length}} \times 100 = \text{C.S.F.}$$ (Crack Susceptibility Factor)

It becomes evident from Fig. 6 that complete freedom from hot cracking might be expected when the C.S.F. has a value of about 20 or less.

Considering the four steels of Table III, crack susceptibility factors derived from the crack test results of Table IV are seen to be 22·6, 18·5, 39·6, and 0.

Whilst the Cr–Mo–V steels Nos. 6 and 7 are seen to possess good resistance to hot cracking, undoubtedly the most significant result is the wide difference in crack sensitivity shown by the air-melted and high purity vacuum melted Ni–Cr–Mo steels to specification SAE.4335 (Nos. 8 and 9).

An earlier paper[4] which investigated the relationship between hot cracking and composition, proposed

a formula by which sensitivity to hot cracking could be assessed. That paper suggested that carbon, sulphur, phosphorus, silicon, and nickel were adverse, while manganese, chromium, molybdenum and vanadium were beneficial in their effect. With the very obvious exception of the high purity vacuum melted steel No. 9, the relative crack sensitivity shown by all the steels tested in the present work is in substantial agreement with the findings of the earlier paper; the effect of relatively high sulphur and phosphorus contents being evident in the RS.120 steels of Table I, the good resistance to cracking shown by the Cr–Mo–V steels Nos. 6 and 7, and the marked crack susceptibility shown by the air melted Ni–Cr–Mo steel No. 8, the latter showing a sensitivity to cracking equivalent to the RS.120 steel No. 5 which gave the highest incidence of cracking during production welding. However, steel No. 9, which is of the same specific composition as steel No. 8, shows zero crack sensitivity. Although there is seen to be an appreciable difference in the sulphur and phosphorus contents of steels 8 and 9, this fact alone cannot, perhaps, explain the wide difference in crack sensitivity shown by the two steels. Therefore, in seeking an explanation of this differing crack susceptibility, other factors must also be considered.

In addition to possessing a higher sulphur and phosphorus content than steel 9, steel 8 contains 0·05% copper, 0·005% tin and 0·021% arsenic. The effect of copper has been noted by Jones[5] who obtained an 'indication' that this element in amounts of the order of 1% increased the hot cracking of low-alloy steels. It would seem from this that the role played by copper is a minor one and, therefore, the small amount present in steel 8 can hardly be significant. The small difference in tin content of the two steels (0·004%) would suggest that this element also is a factor of little significance in their differing crack susceptibility. Of the possible adverse elements (other than sulphur and phosphorus) known to be present in steel 8, the elimination of copper and tin points to its arsenic content (0·021%) as being an important factor in the crack susceptibility of this steel.

There would seem to be little conclusive evidence in the literature on the effect of arsenic on the weldability of low-alloy steels, but a recent B.W.R.A. report,[6] which investigated the effect of this element on the weldability and notch toughness of mild steel, concluded that arsenic up to 0·10% has no effect on the fusion weld crack susceptibility, joint strength, ductility, and notch toughness of mild steel plate, provided that the sulphur and phosphorus contents are within the normal limits of 0·06%. It is well known, however, that such levels of sulphur and phosphorus would be quite intolerable in high-strength weldable low-alloy steels (the latest specification for such a steel gives maximum values of 0·010% S and 0·015% P). It is an interesting possibility that such steels may be equally intolerant of arsenic.

At the present time considerable effort is being expended on the development of new, high-strength steels for rocket motor and pressure vessel manufacture, such steels being required to possess the additional property of good weldability. It would seem that answers to many of the questions concerning the constitution of a high-strength weldable steel lie in the markedly differing crack sensitivity shown by steels 8 and 9.

Conclusions

A simple test has been developed by which the susceptibility of steel sheet to hot cracking may be determined, weldability or hot-crack susceptibility as indicated by the test showing very good correlation with practical welding experience.

A relative measure of a steel's susceptibility to hot cracking, termed the 'Crack Susceptibility Factor', has been derived by expressing the average crack length of the test specimen as a percentage of the test section length (distance between slots).

Use of the test to determine the crack susceptibility of four special steels has produced two results of outstanding significance: the air-melted steel to specification SAE.4335 which gave a C.S.F. of 39·6, indicating marked susceptibility to hot cracking, and the high purity vacuum melted steel to the same specification, which gave a C.S.F. value of zero.

The reason for the wide difference in crack susceptibility shown by these two steels of the same specific composition obviously merits investigation. Such an investigation would, seemingly, provide much valuable information on the constituents that encourage or induce hot cracking in steel during welding.

Acknowledgments

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Autumn Meeting 1960

THE ROLE OF THE METALLURGIST IN FABRICATION INVOLVING WELDING

The Autumn Meeting of the Institute of Welding was held in London from 1st November to 3rd November 1960. The third session of the technical discussions was a Joint Meeting with the Institution of Metallurgists, with Dr N. P. Allen in the Chair. The subject was introduced by H. E. Dixon and M. Birkhead.

Mr. H. E. Dixon (Atomic Power Construction Ltd.): My contribution is intended as an opening gambit to arouse you, rather than a lecture. For some reason it was thought that I was equipped for this because I have been involved in metallurgy for a long time – although I now have a much wider field of interest, primarily from the engineering point of view.

This places me in a somewhat awkward position. I have to look at the subject from several sides: from the point of view of the metallurgist, and from that of the engineer, the designer, and the constructor. I am not quite sure with what voice I am going to speak, so if there is any ambiguity or inconsistency in what I say I would ask you to bear with me.

The very title of the discussion, "The Role of the Metallurgist in Fabrication involving Welding"— in other words, the part that the metallurgist plays in welding fabrication— poses a number of questions.

Firstly, the wording implies that the metallurgist occupies a very special position in relation to welding fabrication which is perhaps different from his position in relation to any other fabrication process. Secondly, does he play the same role in relation to the various industries that employ welding fabrication, or in relation to industries concerned with different materials? Is his role different, for ferrous or non-ferrous, or very special materials? Thirdly, and I think this is the most important, is he playing the correct role? It is on this aspect that we can, of course, have the most discussion. Fourthly, and perhaps it should be the first question, what do we mean by a metallurgist in this particular context? The best metallurgists I know in this particular field are not really metallurgists at all; they are often engineers, chemists, or others who have subsequently acquired a metallurgical knowledge. This raises many questions in regard to training, etc.

For the purpose of this discussion I have assumed that we are interested in the welding metallurgist—a person who, by both training and experience, has specialised in the development of the science of welding. He has concerned himself particularly with the effects of welding on the properties of materials, the properties of weldments as a unit, or the properties of weld metals in particular.

In relation to this question we might ask ourselves very carefully whether we look upon welding itself as a science or an art. In the past it was very much an art. In the last ten or twenty years it has come to be considered very largely as a science, although there is still a little art left in it. This means that the metallurgist in whom we are interested—the welding metallurgist—must have an outlook that enables him to accept both the scientific and the artistic aspects. I think one could lay down immediately the requirement that he must have a proper understanding of the basic principles involved. The welding of steel, fortunately, requires less of the cookery-book qualifications, or trial-and-error methods, and today more and more is being done to evaluate the fundamental principles underlying processes; the effect of processes on materials and the associated physical metallurgy.

A knowledge of basic welding procedures helps a great deal in understanding processes, but much has still to be done on trial-and-error methods. Although we understand the basic principles, we cannot always predict exactly the precise conditions that are required for welding.

The welding metallurgist has to understand the basic principles and must help to further the understanding of those principles. He must also understand the other aspects, which are not all scientific, and must realise that he is often in the hands of the operator. He must understand how to get the best out of welding operators, and how to educate them to understand the things that matter.

Does he really occupy a special position in relation to welding, as compared with other fabrication industries? Welding has grown so much, and is today such an important fabrication process that I find it difficult to come to any other conclusion than that he *does* occupy a different position. He is concerned, for example, not only with materials that should be welded, but with welding and the properties of those materials. He is concerned that the right materials are selected for welding, and that the right techniques are used to get the desired result.

He is consequently much more involved in these matters than, for example, a metallurgist employing some other fabrication process. He is much more concerned in the use of materials. He has, therefore, a very close contact with the user, the engineer, and the designer.

The roles that the metallurgist can play are quite varied; I shall consider two basic ones. He can be primarily concerned in furthering the science of welding, by means of research. He can do this in a research association such as B.W.R.A., in the laboratory of an electrode manufacturer concerned with developing particular materials, or in one of the technical associations or material suppliers, such as a steel works or an aluminium works. He can undertake research as a user as well.

When he is doing this sort of work, where he is primarily concerned with welding research, I consider that it is not

possible to divorce it from the production application or requirement. Here, B.W.R.A. probably does more welding research than anyone else in this country, and although it is engaged in a great deal of fundamental work, most of its staff keep their eye very much on the production applications, or the production problems that arise. In fact, some of their metallurgists are very good engineers, and some of their engineers are very good metallurgists. The combination of these qualifications puts them in an extremely favourable position not only for making progress on the science side but for keeping in touch with what other people are doing and then seeing that the progress is fed to the user who needs it for a particular purpose. The research metallurgists at B.W.R.A. are encouraged to develop the engineering outlook so that they provide an extremely valuable service.

I stress this because, from the user's point of view, the metallurgist who is primarily a metallurgist in relation to welding, and has little other knowledge, greatly limits the field of his activity. I can explain this best, I think, by considering the case of my own particular company and the position that metallurgy plays in it. The chief metallurgist is responsible for the materials section. He is, if you like, a materials engineer—the man who says that, for *this* application, *that* material will be used. He is therefore very much concerned in the environmental conditions in which materials are used, whether these be simple conditions of temperature and pressure or much more complicated conditions involving gases or reactive materials. His word is law on the material that shall be used; no-one can override him on this, so far as a particular application is concerned.

He is, therefore, concerned very much with ensuring that the right materials are being used or specified. He can say "Yes" or "No", and it is as simple as that. In addition, the chief metallurgist and his staff are responsible for fully approving all welding design features, to ensure that the right edge preparations, the right techniques, and so on are being used. You might say, "What has that to do with a metallurgist?" It has a lot to do with a welding metallurgist who is trained in a particular way.

Next, he is also responsible for ensuring that the inspection techniques carried out on the production side are adequate. Although he is not responsible for inspection, he is entirely responsible for ensuring that the right techniques are employed, and the right standards followed; further, that any unusual features are being given special attention. In addition, he is also responsible for ensuring that our research department is doing the research required by the design or construction departments, whether this be on welding or any other process. You might say that he holds an unusual position. I know that, from his point of view, he holds an extremely satisfactory position. I do not for one moment think that many will agree with me that his responsibilities should be as I have outlined them. I am conscious of the fact that it is not possible to be dogmatic about what a metallurgist should do in the field of welding. It depends very much on the people that you have and the training that they have had. Nevertheless, I am putting this forward as an example of a responsibility which, from my point of view, is extremely satisfactory, and here I am speaking from both the design and construction sides.

The more general field of the metallurgist is, of course, as a 'trouble-shooter' in matters involving welding. He is brought in when trouble arises. He is not always brought in at the beginning of a job and given an opportunity to advise whether things are right or wrong. I know of too many metallurgists who are brought into the picture only when trouble looms—though I am sure that this does not apply to any of your organisations! or firms! A major crack occurs and then the metallurgist is in great demand. What often happens is that the solution is found but no one is quite sure why it works. Perhaps six months' later the engineer will get an answer which tells him exactly why and he then assures himself that it merely confirms his own vision in having taken the right decision, even if for the wrong reasons.

This is not a very satisfactory role for a metallurgist but, of course, the role that he has is the one that he really builds himself into. The more help and advice he can give on these matters the surer he is of being brought into the picture at the right stage.

However we look at this we get back to the very personal issue of the role that the individual metallurgist plays which depends very much on himself. Even if he is in the right organisation he may not, because of his own particular limitations, be able to play a proper role. A great deal has been said to the effect that the metallurgist's job would be a great deal easier if the engineer knew more metallurgy and appreciated more of the problems involved. My experience has been that the man I want is the metallurgist who has acquired an appreciation of engineering. Given the right degree of responsibility he can play a much more important part if he has a knowledge of processes, design, and other problems that arise in the course of production. This probably accounts for my metallurgist having the responsibility that he has. It depends, of course, very much on personalities, but I feel that if the metallurgist wants to play a very important and vital role in production he must know the engineering side thoroughly at all levels, including design and production problems. He needs to be a real paragon if he is to become a power in the firm. He has to be all that the welding engineer is. He has to understand the welding and engineering points of view. He has, in fact, to be a welding engineer in embryo, so that he can quite easily step into the engineer's shoes if necessary. He must understand the "electrics" of every process. A little knowledge of mechanical engineering is a good thing as well. He needs to be something of a chemist, and to understand labour relations, and how to put over his ideas to designers, not by wielding a stick, but by education and persuasion.

From all this it will be seen that he has to be a very special person with very special qualifications. I wonder how many of us measure up to those requirements? He must be conversant with codes of practice, and understand practical problems. He must be equally at ease working on the shop floor or with the management, and be in a position to maintain appropriate contacts with his fellow-workers—sometimes the people in the research associations as well—and feed his information through at all levels. Do people such as this exist? I shall not answer that question because, quite frankly, when one starts looking into the role that he might play, and the role that he frequently does play, the second can be seen as not always very important. This brings me to the matter of training and experience.

The metallurgist is trained at a university or a technical college and considers himself qualified, whether he has a degree or a professional qualification. The amount of welding knowledge, even on basic principles, that he receives at either institution is quite limited. He gets his additional knowledge either by joining B.W.R.A.—a most useful source of staff for industry—or a large, progressive company with its own laboratories, where he comes in contact with electrical and mechanical engineers, with physicists and chemists.

Some of the best metallurgists that I know are chemists by training. One of them is a physicist by training. Another is an engineer. In other words, there is no real means of training a metallurgist for welding fabrication except the hard way, through experience. We might well ask ourselves whether the biggest contribution we could make in getting the right role for the metallurgist in welding fabrication would be, in fact, to encourage a more progressive outlook in regard to welding instruction itself, in both universities and technical colleges. For welding, as a process, there are certain technological principles related to processes, which

govern what one can do and cannot do; but the basic problems in terms of welding and the facts concerning materials are admirable subjects for use in training colleges. The physical metallurgy side is extremely good, and covers all aspects of heat treatment, thermal cycling, stress-strain relationships, and so on. One could easily put an emphasis on welding at an early stage in training. On the whole, I believe that is not being done. It could be done by way of post-graduate courses, and I think that we would all like to see more of them. We have one in the country at the moment, perhaps two, but certainly the figure is not impressive.

The metallurgist of whom we are speaking requires a good deal of experience which, at the moment, he can get only by doing the actual job and making his way as best he can. He can gain a good deal of experience by working in a research station which has a forward outlook and is feeding engineering 'know-how' into all its work.

My answer to the first question that I put to you: "Does the metallurgist occupy a special position in relation to welding?" is that I think he does. I think he could occupy a much more important position, according to his greater knowledge of engineering aspects and of the process side as well. The first-class welding metallurgist should have a sound engineering background. It can do no harm and will certainly help a great deal in making him more useful to his firm.

Does the metallurgist play the same role in welding processes in different industries? My feeling is that it is very difficult to be dogmatic about this. Many of the producers of ferrous and non-ferrous materials have progressive policies in their welding departments. These firms, because of their very nature and the fact that they are selling material, are continuously viewing, not merely the narrow field but all of the user aspects. In some works the metallurgist is primarily a trouble-shooter, and is only called upon to get somebody out of a mess; I think that these are in the minority.

Does the metallurgist play a different role according to the type of process? I think that the day has gone when one had a metallurgist who specialised in resistance-welding. I think that I was the last of that line. I do not think that today resistance-welding is regarded as a special field in which particular metallurgical problems are posed. It poses mechanical, electrical, or design problems, all of which can be solved by the right metallurgist, of course, but there are so few in that category that one can almost leave them out of consideration.

The metallurgist who goes furthest these days is the man who knows most about the greatest number of materials, processes, and the engineering and design aspects. This is very hard to dispute. He can do his job so much more effectively than his fellows. Is he playing the correct role? I do not think that one can refer to a definite, fixed role. It must have relation to the individual. Each can answer it himself by asking whether he feels that he is in fact earning his money.

Mr. M. Birkhead (Mather and Platt Ltd.): Our welding industry is extremely varied in its requirements. A small works producing mild steel fabrications to an average standard represents one particular requirement. Going up the scale, increasing the organisation in size and scope, presents very different problems. The small to medium-sized fabricating company dealing with a wide variety of materials represents a field where the requirement is for men who not only understand materials but, more important, are able to help the company produce at a certain rate. They must earn their keep, and must move things from A to B quickly, and that in the correct manner.

Mr. Dixon pointed out that the metallurgist is often brought in as a 'trouble-shooter' in this sort of company. If that happens it is much too late and the metallurgist is in a hopeless position; but if we take a man who is trained basically as a metallurgist and then, by environment and study, gains a practical understanding of welding, electrodes, processes, and the design side, he can be a great asset in the early days of a fabrication, before it goes through as a design. He can embody his own ideas and carefully follow them through in a department for which he is responsible as metallurgist. He is virtually the welding engineer. This can probably apply only in a small organisation, but if one can find such an "animal" one may consider him a specialist.

In the larger company, I feel he should advise on all matters concerning metallurgy, both general and welding—having in mind the diversified interests that are encountered today. There is a need for someone who can say, "This is the correct material to use. This is the way to do it". When the specification is in writing his responsibility ends, in the sense that the design will be produced and followed through by the production staff. He may find that the welding engineers, or the production staff, have done something which is not necessarily connected with the particular specification. I still think that in the larger organisation I would prefer to see the metallurgist in a role in which he can not only advise but be able to specify something that will be carried out.

I feel that there is at present a distinct shortage of technologists who can combine metallurgy and welding. I prefer the combination of metallurgist-welding engineer rather than engineer-welding engineer. I would like to hear engineers' views on these matters.

Mr. Dixon touched very briefly on the question of enlightened management. Welding is a comparatively new industry, and in many places welding is viewed as a method whereby one joins two pieces together, and as long as they stay together everyone is satisfied; if not, the metallurgist is brought in. With a lead from management, metallurgists could be employed as they should be employed, with very happy results.

Mr. Dixon mentioned the British Welding Research Association and the other bodies that provide scientific information for the industry and mentioned that the results of research must be put into practice. No one is better fitted to do that than the metallurgist; he can see that the formination reaches the design departments and the production departments, and thus fill his role of controlling and acting in an advisory capacity, with authority.

The metallurgist who has some knowledge of practical production will gradually co-operate in the more theoretical aspects of design, so that it will be possible to make the product that is fit for its purpose.

Mr. R. G. Braithwaite (Past-President): I cannot help reflecting how much we owe to the fact that the pioneers who developed the process of welding had a metallurgical background. Metallurgists have since played an important part in welding, but they are only part of the team, and metallurgy represents only one aspect of the training of a welding engineer. He must be something of a chemist, and have an engineering and design background, but basically he must depend upon metallurgical knowledge. We have to work with metals, and weld them. The metallurgist has to produce the weldable materials. He must therefore work with us and understand our problems.

Mr. J. Hinde (International Nickel Co. (Mond) Ltd.): The metallurgist's most important role in welding is to put his message across to the welder; this requires close contact with the men on the shop floor. Yesterday the President showed some samples of defective mild steel tubes. These had been joined by competent welders but microsections illustrated the defects. The cause of the trouble was really a change in the thickness of the tubing from $\frac{1}{16}$ in. to $\frac{1}{8}$ in.; the welders had worked on $\frac{1}{16}$ in. material for so long and

used the same technique on the thicker material and this was the cause of the trouble. Apparently a talk with the welders soon put matters right but it is my point that, without such contact, the metallurgist could have wasted a great deal of time in the laboratory assessing the nature of the defect and not helping anyone very much.

The metallurgist must be in a position to go into the shops and decide for himself whether a particular problem is metallurgical or the remedy lies more in attention to the welder's piecework rate! The metallurgist needs experience to do that; experience that cannot be acquired in the laboratory but only amongst the welders on the shop floor. Although it will be difficult for the metallurgist to match the skill of the practised welder, it will be a great advantage if he can put down a weld. The metallurgist must get his information to the shop floor where it can be put to most use and this difficulty can better be overcome by personal contact than by the submission of reports.

This afternoon the effect of phosphorus on the hot cracking of steel was mentioned. One solution of this problem is to use a steel with only 0·001 % phosphorus but costing around £1,000 a ton and the price for matching weld-metal would be about the same. One wonders what the cost of the final weldment would be and I would like to take this subject to illustrate another feature of importance to the metallurgist concerned with welding. The metallurgist should be aware that the immediate solution of a problem is not necessarily the correct metallurgical answer. Often he may have to compromise because of the cost involved or some other factor, like expedience of delivery of a particular material. He cannot do this, however, unless he has a wide knowledge of all the aspects involved, including the many that have been mentioned by Mr. Dixon and Mr. Birkhead. The metallurgist will sometimes have to be reconciled to not having his metallurgical advice readily accepted, but he must be aware of all the factors involved before he can compromise to the advantage of all concerned or maintain his own point of view against all opposition.

Those laboratories where new materials are developed may not be in quite the same position to maintain contact with welders at what I have called shop floor level. In such circumstances the metallurgist must have the courage sometimes to say a material is unweldable by this or that process. His assessments will be based mainly on test pieces but I would like to emphasise that whatever may be done in the laboratory, ultimately someone must make a component. Assessment will be speeded up by making a dummy part from the new material which one might think will serve the purpose and in this connection, the research metallurgist will do well to discuss the matter with those metallurgists who have the shop floor experience.

Mr. N. T. Burgess (C.E.G.B., Research and Development Dept.): Welding is becoming more and more closely linked with inspection, whether destructive or non-destructive, and I find it increasingly difficult to decide whether the people responsible for welding should be responsible for acceptance or rejection. I should like to hear the views of expert welding engineers on where the metallurgist, as distinct from the welding engineer and the designer of the component, fits into weld inspection.

Mr. J. A. McWilliam (Firth-Vickers Stainless Steels Ltd.): Metallurgists played a most important part in solving the problem of hot cracking in 18% Cr–12% Ni–1% Nb welded steam pipes, by the use of 17% Cr–10% Ni– 3% Mo pipes.

Mr. Fitch raised the question of making 18% Cr–8% Ni electrodes using mild steel core wire and having all the alloys in the coating. This was tried thirty-five years ago and the resultant electrodes were unreliable. The length of the arc is, of course, an important feature. Mr. Fitch mentioned that this could now be done with touch-welding electrodes, but even the best welder does not always keep the touch all the time. I do not think that this is the complete answer.

Mr. I. C. Fitch (A.E.I. Ltd.): I shall put forward a heresy: I say that a welding engineer is an anomaly; that his function is really that of a process engineer. He has to deal with so many different processes and so many different methods of attaching materials that he cannot hope, if he is to do the job properly, to cover the whole field.

I take the view that the metallurgist must be at the elbow of the process engineer of whom I have just spoken, to advise him on the problems that arise from the metallurgy of welding. An organisation may not always be big enough to justify having two people, but fundamentally this is the correct approach because the metallurgist needs time to study the problems that cannot readily be answered by the process engineer, free from pressure of daily production problems.

In answer to Mr. McWilliam, the stainless-steel electrode made with mild-steel core was not one that I would recommend for all purposes; but there were certain applications where a little extra tolerance involved no great risk, and under those circumstances an economic advantage could be gained. I quoted it as an example of what could be done, and has indeed been done for a very long time.

Mr. S. W. Neville (A.P.V. Co. Ltd.): It is not surprising that the general trend of this discussion has been towards the appraisal of the metallurgist, but I think perhaps a word should be said on behalf of the engineer.

The metallurgical graduate probably sees a better prospect in welding technology than does his engineering counterpart. He sees his chance of studying molten metal neatly contained in discreet little pools, with the additional opportunity of using his basic engineering instinct for plotting cooling curves, calculating deposition rates, and the like. Thus he can say that he will be doing an engineers job as well as that of a metallurgist, even to such an extent as to adopt the title of "welding engineer"!

The attraction that welding offers to the mechanical or electrical engineer is not nearly as strong, probably because he has such a vast field in which to choose his path. The manufacture of motor-cars, prams, tin pots or wireless sets, is open to him on every side, and welding is a comparatively narrow gateway. Once inside he is further confused by the baffling array of metallurgical engineers, engineering metallurgists and the rest!

The fact remains, however, that we need metallurgists, and we must have engineers working on welding, and each has an equal place. It depends very much on the personality and ability of the individual whether he will reach one of the exalted positions previously mentioned, and he then can call himself whatever he wishes.

Basic welding research, such as is undertaken at B.W.R.A. essentially calls for a metallurgical approach. The engineers are the men who will take the welding processes, as they are offered by the manufacturers, select them, and install them on the job. They will plan the work to suit the process and will be responsible for the quality of the finished product.

There are, of course, the geniuses whose light shines in both fields but they, after all, have little need to concern themselves with a discussion such as this!

Mr. W. J. Colton (Newton Chambers and Co. Ltd.): I agree with Mr. Dixon that the works metallurgist must be a man of many coats, or hats, but certainly he cannot be merely a 'straight' metallurgist, trained in metallurgy and that alone; he must have a very strong engineering bias. I do not think it matters much whether one begins as an engineer and then becomes a metallurgist, by whatever

means are available, or *vice versa*. I do think that it is essential that one should be able to appreciate the engineer's point of view, speak in his language, and understand his side of the matter.

I am not sure that I should want engineers to have to accept designs from metallurgists. Certainly, it is for the metallurgist to specify the way in which things shall be done and to deal with such important matters as stress distribution, but design is not his province.

I feel that the metallurgist should state the how and when of inspection but should not be responsible for the inspecting function. He is then getting rather outside his proper sphere, and in the larger organisation his main role should be advisory, rather than on the shop floor.

Fortunately the practice of calling in the metallurgist only when things go wrong is dying out. People are realising that they can be helped before mistakes are made, and are taking advantage of this fact. On the other hand, the metallurgist is often sandwiched between designers on the one side, and the engineers and the shop floor on the other, and is also the contact between them. He can help them both.

As Mr. Dixon has pointed out, the role that the metallurgist plays depends largely upon himself—I was almost tempted to say, "irrespective of his training", assuming that it is basic. Whatever the combination of his qualities may be, his success will ultimately depend upon himself, and how effective he is in giving the correct advice and establishing good relationships with people on his own level who are specially concerned with engineering, design, and the shop itself.

Mr. P. M. Bartle (B.W.R.A.): What I should really like to talk about is how one becomes a good metallurgist. Various speakers, especially Mr. Dixon, have told us that they know many engineers and chemists who are also good metallurgists; but the question is how to join that category. Most of these people have had probably twenty years' experience and have learned their metallurgy the hard way. Are we going to find our men this way, or train them as metallurgists from the beginning? And if we adopt the second course what is their job to be?

For a metallurgist to be able to do a useful job he should have a thorough grounding in metallurgy and a knowledge, but not an intimate one, of engineering, chemistry, corrosion, and other subjects with which he is likely to come into contact. As a research metallurgist connected with welding, I often wonder how much use it is to have a thorough knowledge of the welding process. It is relatively easy to learn what one can or cannot do in welding, but, in my opinion, the welding metallurgist must be prepared to try out the way in which something should work and should not work, and see if it can be *made* to work. It is probably far more important for people developing actual processes, as opposed to applying processes or improving them to have a sound knowledge of the requirements, but I still think that they must be ready to try the wrong thing as well as the right thing.

I thoroughly agree with Mr. Birkhead's remark, that the work done by the metallurgist changes greatly as he goes up the scale; his outlook must also change. Mr. Birkhead also made the point that the man who is going to be a welding metallurgist should be able to weld and handle an electrode. I agree that this is probably a good thing, but I would point out that there are many very good welding metallurgists who are probably very bad welders.

Mr. C. L. Railton (T.I. Technological Centre): For want of something better I shall call myself a welding engineer with a smattering of metallurgy, and that seems to me to be one of the main difficulties.

I heartily agree with Mr. Dixon that the ideal man, whether he starts off as an engineer or as a metallurgist, is of the kind that he has described; but all too often the welding engineer, or the production engineer, gets into trouble and, upon asking the metallurgist what is wrong, receives an answer in a language that he cannot understand. That is where rectification is needed. I have, unfortunately, had to deal with general graduate metallurgists who looked on the welding engineer as the man who deals with messy jobs—certainly nothing so refined and delicate as looking at a beautifully polished specimen under a nice, expensive microscope. If one asked them to do a welding job, or something of the kind, they would dash through that part of it as quickly as possible so that they could wash their hands and get a nice, clean specimen to look at under their microscopes—then giving one the answer in language that one could not understand.

An earlier speaker said that welding experience was an excellent thing for training the metallurgist. In resistance-welding one might ask the metallurgist for the *S* curves for quench rates occurring in times of perhaps 5 cycles—(though the metallurgist would not really understand because, to me, that is 0·1 sec.)—and would probably be told that there were no such curves. Nevertheless, the poor welding engineer needs information of this kind. Welding therefore, seems to offer excellent scope for training and to be a fundamental need for the young metallurgist, whether he is going into a steelworks or the welding industry. I feel that graduates or post-graduate students in metallurgy should be given far more information about welding as well as experience of welding processes. A metallurgist may say that a material is easily weldable, but if I answer that I am going to quench it in 0·4 sec. he will say that it cannot be done. My answer would be that it could be done by resistance-welding, but he will probably never have heard of this. The metallurgists need to realise what dreadful shocks welding can give to the materials that they know so much about.

I should like the metallurgist to be able to translate his knowledge into terms that the ordinary engineer can understand—to say what it means in rates of heating and cooling. If not, the welding engineer, or the ordinary engineer, has to learn something about metallurgy. We have today been told of heat-treatments; of sixteen hours of this and ten hours of that. This does not help me very much if I want to flash-weld an engine ring. I want to know whether the setting is right *now*—not in sixteen hours' time—because there are a hundred other rings waiting to be done. The metallurgist shakes his head and likes to go away and think about it. The welding engineer probably decides to try bending it at various angles until he finds out where breaking-point is reached. He will probably heat-treat a sample for the required sixteen hours and later see whether there is any correlation between the results, but he cannot wait that long before going on with his work.

Mr. J. V. Thomas: Through many years of experience in engineering I have come to the conclusion that "consistency" is furthest from our thoughts in the works. I should like to give the metallurgist the very real job of getting his own systems consistent, so that when one metallurgist examines a section he comes to the same conclusion as his colleague next door. I have dealt with four or five metallurgists at the same time, and one simply cannot find consistency in their results.

Let us get down to real fact. We have been welding mild steel for fifty years and stainless steel for thirty-five. We have been welding alloys for a comparatively short time. Poor Dr. Cottrell has been welding a very difficult alloy for the last seven years and is beginning to make some progress. We are coming to see what a weldable material is. The metallurgists are very welcome to come into the shop and watch the processes but then I hope that they will go away and get us weldable material.

I do not care what the price is; if you can get a welding material that will produce a reasonable finish, coupled with simplicity, you will be making progress. In some industries a great deal of material is scrapped and left in the stores because it could not be welded to the required specification. If the metallurgists can help they are more than welcome because there are more problems than we care to think about on the fabrication side, due mainly to the materials that one gets.

Mr. Dixon said that his metallurgist specified the material to be purchased for a given job and checked it when it came into the shop. I should like to ask him how much he sends back. Can one get a material to specification? The specification is for manganese 0·8 to 1·9% for instance. That is ridiculous. Dr. Cottrell knows what happens; the carbon content is 0·1% at one corner and 0·4% at the other. How can one weld a sheet like that? The metallurgist has a difficult job. I would ask him to get suitable material for us to weld. Perhaps he should then keep out of the shop.

The President (Mr. E. Fuchs): Mr. Railton touched on a key matter when he spoke about "language". In welding, the important thing is to have a team, such as one finds in Mr. Fitch's firm, and in our own, of engineers and metallurgists speaking each other's language or, in smaller firms, to have one welding technologist.

We are striving very hard in the Institute and the B.W.R.A. to bring welding education up to date, for it is a long way behind. I would ask anyone who sees a movement towards better welding education in any of the colleges of technology to foster it to the greatest possible extent because it has been shown that the people who are behind in this matter are often not in the colleges but in industry which has to provide or demand students of welding technology.

This autumn there will be established at the College of Aeronautics, the Ministry of Education's own college, a three term post-graduate welding course that will welcome both engineers and metallurgists and will be so designed that they will be able to learn each other's language as it applies to welding.

In Scotland, Mr. Barr, one of the Vice-Presidents, is working extremely hard to introduce in the metallurgist's H.N.C., an endorsement in welding very much on the engineering side, and to introduce into the engineering H.N.C. an endorsement very much on the metallurgical side. For these endorsements the metallurgist would learn a little engineering language, and *vice versa*, I would ask you to do everything you can to encourage this sort of thing in your own areas and we shall also do so. Do not leave it to the educationalists; speak to your managing director.

Mr. H. E. Dixon: I agree entirely with what Mr. Birkhead said. Mr. Braithwaite and Mr. Thomas, mentioned the desire to produce weldable materials. That is a very good thing, but I must point out that one can get things as long as one is prepared to pay for them; I can obtain much better mild steel if I am prepared to pay for it. The magic word is "economics". We must, however, remember that generally speaking we have to use things that we can pay for. If I want a sulphur limit of 0·03% and a phosphorus content of 0·001% I can get it, but the price will be rather shocking!

I do not think that there should be any doubt in any one's mind that the metallurgical industry has gone a long way in developing materials and meeting welding requirements, but it is obviously not possible to satisfy all engineers. It is not possible for me to be satisfied in my present role, bearing in mind what I would like to have. The fact is that some things can be achieved and some cannot.

I agree with Mr. Hinde that contact with the shop floor is most important. I do not know whether Mr. Bates was suggesting that I had said something wrong, but I really did draw attention to the need for a knowledge of what went on on the workshop floor, and the desirability of being able to put things across. It is necessary to sell oneself upwards and downwards, as it were.

I did not specify that the metallurgist should be responsible for inspection; I said that he should be responsible for laying down what was important, what was not important, and what should be looked for. Physical techniques should obviously be in the hands of the experienced physicist.

Mr. Neville mentioned graduates, and how they found their way into welding. I should like to know how many have come into welding in the last few years, whether on the process or metallurgical side. My experience is that most graduates today seem to be possessed by the idea that there is a branch of physical metallurgy, or near-physics, called solid-state physics which is the be-all and end-all, and I find it very difficult to discover metallurgists fresh from the university who are interested in learning something about welding. I notice with interest a trend in university schools towards dislocation theory and solid-state physics and a great reduction in the number of metallurgists going into the industry as graduates.

Mr. Railton was not sure what he should be called. If he does not know I cannot tell him, but I agree that we all need to broaden our knowledge. Once the trend towards specialisation is followed it is difficult to throw it off and broaden one's outlook. This is true of everyone. There is a tendency to be tagged very early, as a result of something that one has done, and it is difficult to lose that tag; if a man can lose it in twenty years he is doing well. Because of the way we work these days a specialist becomes valuable, but from his own point of view it must be a deadly limitation. We should follow the example of our naval friends, who train all their engineers with the intention of making them Directors of Naval Construction, although only one person at a time can ever occupy that position. It must nevertheless be a very satisfying training. With a broader outlook we should all become potential managing directors, and in my view this would not be such a bad thing.

Welding RESEARCH

B.W.R.A. REPORT

Suggested Explanation of Hot Cracking in Mild and Low Alloy Steel Welds

By J. C. Borland, A.I.M.

An explanation of the effect of composition on hot cracking is given for binary iron alloys, mild and low alloy steels using in part the "Generalised theory of super-solidus cracking" developed previously by the author, and published literature.

Using a relative Potency Factor $\left[M_L\left(\frac{1-K}{K}\right)\right]$ *it is suggested that the most harmful elements by themselves, in decreasing order of potency, are S, B, (Ce?). P, As, Nb, Sn, Zr, and Ta, and they are probably much more harmful in alloys initially solidifying as austenite than those freezing first as ferrite.*

Oxygen, and deoxidisers such as silicon and aluminium, can play an important part in changing the distribution of the liquid phase, and thereby altering the cracking tendency.

Introduction

DESPITE the large number of investigations on the hot cracking (super-solidus) propensity of welds, the fundamental reasons for cracking are still not yet completely understood. It has been established, however, that the main factors causing cracking are mechanical and metallurgical. The mechanical factors include size and thickness of material, joint design, and size and shape of the weld bead. In general the greater the strains imposed on the weld during solidification the more severe the cracking. The principal metallurgical factor is the presence of a liquid phase at the grain boundaries during freezing. What is not known, however, is the exact nature of the grain boundary liquid and in what manner it acts to cause cracking.

A "generalised theory of super-solidus cracking",[1] recently developed by the author, attempts to explain the fundamental reasons for cracking in welds (and castings). This present paper discusses the applicability of this theory (and its extension) to hot cracking in mild and low alloy steels* and it is hoped that the approach to the problem, developed in this article, in spite of its speculative element, will provide a logical background for future investigations into the development of filler materials for welding high strength ferritic steels. The effect of elements in high alloy steels will be discussed in a future article.

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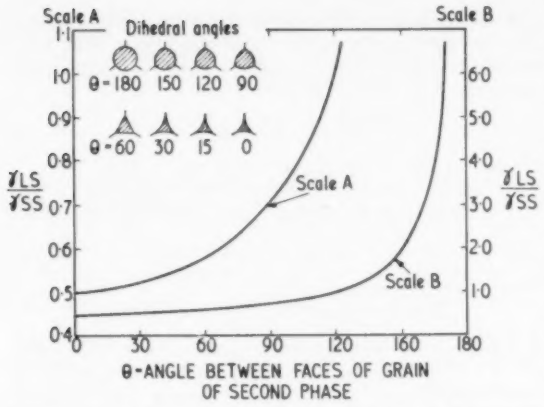
Theory of Cracking

Initial development of generalised theory

This theory explains that for cracking to occur it is not a sufficient condition that a wide freezing range exists; liquid should also be present over a relatively

* W. K. B. Marshall in independent work has also recently surmised on the effect of composition on cracking. (B.W.J. July 1960.)



1—*Relationship between dihedral angle and ratio of interfacial energies (after C. S. Smith)*

wide temperature interval in a distribution which will allow high stresses to be built up between grains. A liquid phase covering *almost all* the grain faces during freezing will allow the development of high stresses on the narrow bridges joining adjacent grains and thus could give rise to fissuring, whereas a liquid confined mainly to grain edges and corners will allow large areas of the grain faces to unite thereby forming strong bonds which are able to withstand the weld cooling strains without breaking. It is further suggested that the liquid distribution during freezing is governed by the ratio τ* of the interphase (solid–liquid) and grain boundary energies. A low ratio (τ slightly greater than 0·5) is extremely harmful as *almost continuous* films are formed; a high ratio (τ=0·57) is beneficial as the liquid is then restricted to grain edges and corners. The appearance and distribution of the liquid phase (when solid) can be seen on a metallographic section, and the dihedral angle (θ) which is a measure of τ, can be determined.[2,3] (See Fig. 1). Since the dihedral angle is a more practical and convenient measure of the distribution of the liquid phase than is τ this will be referred to in the subsequent discussions.

As solidification proceeds θ increases† and, in generalising, it is suggested that the increasing value of θ with cooling, competes with the increasing thermal strains and that fissuring is only prevented when the stresses imposed on the bridges between grains do not exceed a critical amount. Cracking occurs late in the solidification cycle, in a 'critical solidification range' (C.S.R.) where fissures that are developed are not able to be 'healed' because the advanced development of the grains prevents the free passage of liquid (Fig. 2).

To apply the generalised theory rigorously, it would be necessary to know the θ/temperature relationship and the mode of freezing under welding (non-equilibrium) conditions. Information on the θ/temperature relationship is not available so far as the author is aware but some results[2–5] have been obtained under equilibrium conditions and are useful pointers. The freezing mode depends to a very large extent on the solute concentration and welding conditions. As the solute concentration is increased the dendritic growth form gives way to equiaxed crystals. Likewise as the weld cooling rate increases and more solute is rejected ahead of the advancing solid/liquid interface there is a greater tendency for equiaxed crystals to form. This type of growth, in certain circumstances, appears to be less crack sensitive than the columnar mode.

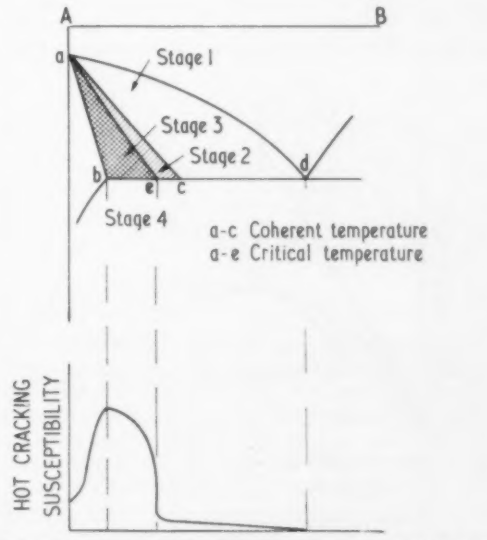
Although data on θ is lacking, the application of the generalised theory need not be too restricted. For instance, it is thought that the value of θ is largely determined by the composition of liquid in contact with the growing crystals; when solid and liquid compositions are similar, small values of θ (near 0°) are favoured; when they are totally different then large values develop. Therefore, systems which to a large extent satisfy the conditions for cracking are those which exhibit a small primary solid solution range, a liquidus which *initially* does not change too rapidly with a fall in temperature, and a relatively wide freezing range. It is assumed, of course, that the extent of the freezing range is related to the extent of the C.S.R. where cracking occurs. The function of the solute is to lower the freezing temperature of the grain boundary liquid, but it is the solvent rich liquid, which in this case is iron, that is responsible for the low dihedral angle developed.

To prevent cracking in binary alloys, one solution would be to reduce the solute concentration to a safe level. Another solution, but much less practical in many cases,‡ would be to add more solute so as to

* $\tau = \frac{yLS}{ySS} = \frac{1}{2\ \mathrm{Cos}\ \frac{\theta}{2}}$ where yLS = liquid-solid interfacial energy, ySS = grain boundary energy, θ = dihedral angle

† The 'effective' value of θ is probably established during the first few moments of freezing, as with the high cooling rates in welds, equilibrium values of θ would only be established at the very high temperatures.

‡ Aluminium alloys are an exception to this rule.



2—*Effect of constitutional features on cracking susceptibility in binary systems (Generalised theory)*

create the condition where large quantities of liquid are freezing over a narrower temperature range. A third solution, predicted by the generalised theory, involves substantially changing the composition of the liquid phase (without greatly affecting the quantity) at the grain boundaries by adding another element which is much more soluble in the liquid than in the solid. This should increase θ.

It is probable that the generalised theory has its main application in ternary and more complex systems.

Effect of the freezing range

It has been pointed out by several authors that as the extent of the freezing range increases so does the severity of cracking for many simple binary alloys. For more complex alloys this rule does not hold, and in many instances this is probably because of the greater importance of the dihedral angle. For solid solution alloys, the dihedral angle is probably just above 0°, which is the condition most likely to cause cracking, and it seems reasonable to expect that the propagation (extent) of the cracks should be related, in some way, to the temperature range in which liquid still persists.

Up to the limit of solid solubility in binary alloys it is assumed that the extent of the C.S.R. will increase with an increase in the freezing range. Alloys with sufficiently long freezing ranges should be prone to cracking. Under equilibrium conditions the extent of the freezing range (ΔT_f) is given by:

$$\Delta T_f = M_L\, Co\left(\frac{1-K}{K}\right) \quad \ldots\ldots\ldots\ldots \quad (1)$$

where Co = solute concentration
M_L = slope of liquidus °C/wt%
K = partition coefficient

The maximum extent of the freezing range occurs when Co reaches the solid solubility limit.

In some non-welding experiments on dilute alloys, Tiller *et al*[6] have shown that the smooth face structure, typical of pure metals, breaks down when the temperature gradient G is less than the gradient of the liquidus temperature (constitutional super-cooling). In other words, non-equilibrium freezing occurs and a homogeneous solid solution structure is not formed. It is probably only under these conditions that weld metal cracking normally takes place.[1] Constitutional super-cooling (Fig. 3) occurs when[6]:

$$\frac{G.D_L}{R} < M_L\, Co\left(\frac{1-K}{K}\right)^{*} \quad \ldots\ldots\ldots\ldots \quad (2)$$

where R = rate of crystal growth, cm/sec.
D_L = Diffusion coefficient of solute in liquid, cm²/sec.

The maximum amount of super-cooling is given by[7]

$$\Delta T_s = \Delta T_f\left(1-\exp\frac{RX}{D_L}\right) - GX \quad \ldots\ldots\ldots \quad (3)$$

where ΔT_s = super-cooling °C.
X = distance ahead of solid/liquid interface.

The amount actually occurring depends very much on the efficiency of the primary structure in reducing super-cooling.

The crystal growth rate R will vary from zero at the fusion line to a rate equal to the welding velocity V at the weld centre line and is given by[8]

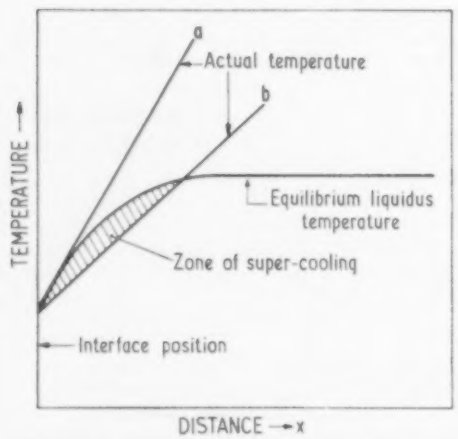
$$R = V \sin\psi \quad \ldots\ldots\ldots\ldots\ldots \quad (4)$$

where ψ = angle between the tangent to the solidus isotherm and the line of welding.

The temperature gradient G in the liquid will be largely determined by the heat input (amps × volts), welding velocity and the thermal properties of the material. In general, the greater the welding speed the *less* G will be at the weld centre line. The gradient between the centre line and the fusion boundary will stay more or less constant. Assuming $D_L = 5 \times 10^{-5}$ cm/sec and $V = 0{\cdot}25$ cm/sec (6 in/min) then at a level of 0·01% Al super-cooling will occur when G is less than 75°C./cm. At 1% Al G would have to be more than 7500°C./cm to prevent super-cooling. With 0·01% P, G would have to be 6000°C./cm. It is easy, therefore, to understand how under welding conditions, non-equilibrium conditions are obtained. Weld microstructure, cracking, and other weld properties will presumably depend on the $\frac{G}{R}$ value.

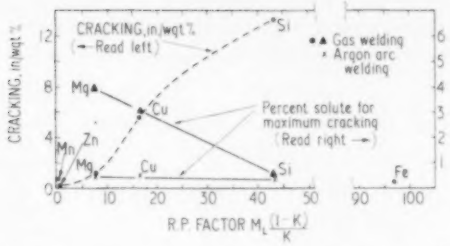
The effect of this factor on cracking does not appear to have been studied experimentally in this way before and therefore it is not appropriate to comment in much detail on its particular effects. However in view of the known effect of the extent of the freezing range it is possible to describe simple relationships between this and cracking, for materials having similar thermal properties, and welded under similar conditions.

The relationship between the extent of the equilibrium freezing range (eq.(1)) and the severity of cracking in some binary aluminium alloys has been discussed briefly by the author[9] using cracking data obtained by Pumphrey and Lyons.[10] In Fig. 4 cracking (in./wt%) is plotted against $M_L\frac{(1-K)}{K}$ which has been called the Relative Potency factor; the bigger this factor the more severe the cracking per unit weight % of solute. The actual amount of cracking occurring is estimated by multiplying the R.P.F. cracking value by the solute concentration Co, though this arithmetical process cannot be extended too far, as the maximum



3—(a) *Equilibrium freezing;* (b) *constitutional super-cooling*

* More usually expressed as $\frac{G}{R} < \frac{M_L\, Co}{D_L}\left(\frac{1-K}{K}\right)$



4—*Effect on cracking in aluminium alloys of factor* $M_L\left(\frac{1-K}{K}\right)$ *of various elements*

cracking that occurs is reached at solute concentrations far below the solid solubility limit, and corresponds to the point where eutectic is first observed.

The maximum cracking (peak value – *see* Fig. 2) obtained is dictated to a large extent by the factors in eq. (3), and to a first approximation it is to be expected that the peak value solute amount will be lower, the greater the R.P. value. Further, decreasing $\frac{G}{R}$ should also decrease the amount of solute required to obtain the maximum cracking as indeed happens when changing from gas to arc welding (Fig. 4). This however is far from the complete story because equation 3 ignores the mechanical strain factor imposed by the welding process.

Normally, to be judged harmful, an element must be capable of attaining a certain minimum value of maximum potency, that $M_L C_S \frac{(1-K)}{K}$ (where C_S = maximum solid solubility). In other words unless the freezing range is sufficiently long, cracking will not normally occur, no matter the value of the R.P.F.* This is the reason for the apparent anomaly of iron in Fig. 4.

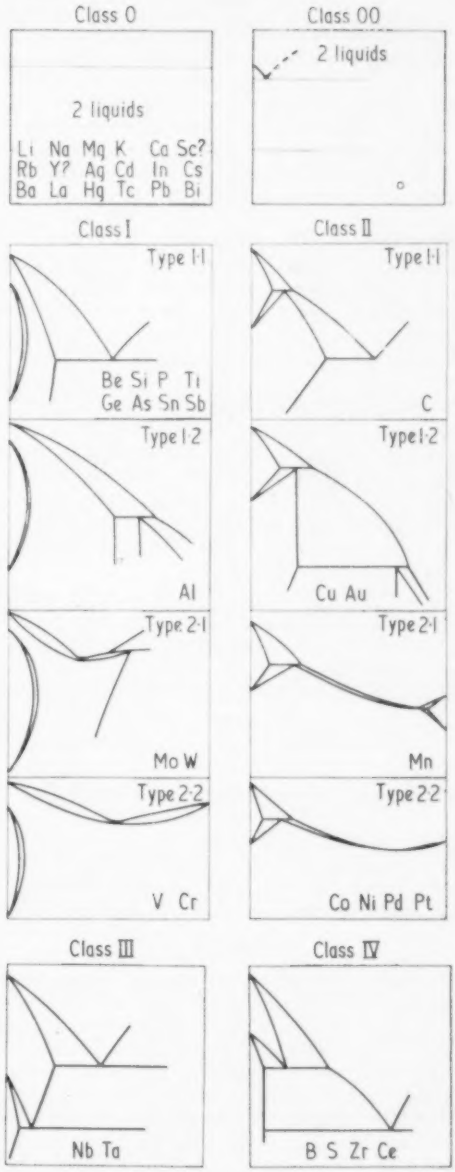
It seems likely that some of these above arguments can be extended to iron-binary systems, and to a less extent to mild and low alloy steels.

Cracking in Iron-Binary Systems

Under this heading an attempt is made to present an idealised picture of the effect of constitutional features on cracking tendencies in binary alloys. The arguments put forward should allow the effects of elements, in more complex systems, to be more clearly understood.

Constitutional features

The element iron is characterised by its ability to exist either in the body centred cubic (alpha) or face centred cubic (gamma) form, depending on the temperature. The addition of a second element tends to stabilise either the alpha, or the gamma phase. Since many of the elements form similar binary systems with iron they may be separated into groups.† The classification system devised here shows, in particular, the relationship of solid and liquid phases. The binary diagram types are shown in Fig. 5 and the maximum solubilities of elements in delta δ, gamma γ and alpha α iron are summarised in Table I. Delta iron is, of course, the high temperature form of alpha iron.



5—*Iron binary types*

* This statement applies to normal welding conditions. It is of course possible to detect an effect due to the extent of the freezing range even when very small, providing suitable tests are made.

† See, for example, K. W. Andrews "Principles involved in Formation of Iron Alloys", *Metal Treatment & Drop Forging*, 1952, vol. 19, pp. 425–432, 489–498.

Table I

Solubilities of elements in iron-binary systems

Class 'OO'

Element	δ Atomic %	δ Weight %	Temp. °C
O	(0·01–0·02)	(0·003–0·007)	1523

Class I

Element	δ Atomic %	δ Weight %	Temp., °C.	γ* Atomic %	γ* Weight %	Temp., °C.
Type 1·1						
Be	33	7·4	1165	3·0	0·5	1050
Si	31·2	18·5	1200	3·5	1·8	1170
P	4·9	2·8	1050	0·45	0·25	1150
Ti	8	7	1340	0·7	0·6	1150
Ge	(19)	(23)	1125	(2·5)	(2·0)	1150
As	(8·4)	11	835	2·8	3·75	1150
Sn	8·3	16·1	1130	1–1·3	2–2·5	1150
Sb	(19–20)	(34–35)	1002	0·9	2·0	1150
Type 1·2						
Al	52·8	35	1232	1·2	0·6	1150
Type 2·1						
Mo	26·0	37·5	1450	1·6	2·7	1150
W	13	33	1540	1·0	3·2	1150
Type 2·2						
V	continuous		—	1·6	1·5	1150
Cr	continuous		—	13·3	12·5	1000

Class II

Element	δ Atomic %	δ Weight %	Temp., °C.	γ Atomic %	γ Weight %	Temp., °C.
Type 1·1						
Cu	(5·5)	(6·25)	1484	7·5	8·5	1094
Au	(3·0)	(10·0)	1430	8	23·5	1168
Type 2·1						
Mn	1·0	1·0	1504	continuous		
Type 2·2						
Co	16·5	17·3	1499	continuous		
Ni	3·24	3·4	1512	continuous		
Pd	3·39	6·5	1478	continuous		
Pt	2·3	7·6	1519	continuous		

Class III

Element	δ Atomic %	δ Weight %	Temp., °C.	γ* Atomic %	γ* Weight %	Temp., °C.
Nb	2·75	4·5	1360	1·2	2	1220
Ta	2·3	7	1410	0·95	3	1220

Class IV

Element	δ Atomic %	δ Weight %	Temp., °C.	γ Atomic %	γ Weight %	Temp., °C.
B	0·75	0·15	1381	0·1	0·021	1149
S	0·31	0·18	1365	0·023	0·013	988
Ce	?	?	?	0·16	0·4	1060
Zr	4·5	7·2	1335	0·5	0·8	1330

* Liquidus not involved. (*See* Fig. 5).

Table II

Cracking potency* of elements in iron-binary systems

Class	Element	R.P. Factor $M_L\left(\frac{1-K}{K}\right)$	$M_L\ Co.\left(\frac{1-K}{K}\right)$ Co = 0·01 %	Co = Max. solid solubility
I	Be	5·41	0·054	40·0
	Si	1·75	0·017	32·4
	P	121·1	1·21	338·9
	Ti	13·8	0·138	97·0
	As	40·2	0·402	442·6
	Sn	16·8	0·168	270·7
	Sb	(5·04)	(0·050)	(176)
	Al	1·52	0·015	53·3
II	C	322	3·22	32·2
	Cu	3·61	0·036	22·5
	Mn	26·2	0·262	26·2
	Co	0·30	0·003	5·2
	Ni	2·93	0·029	18·2
III	Nb	28·8	0·288	130·0
	Ta	13·0	0·130	91·2
IV	B	917	9·17	137·6
	S	925	9·25	166·5
	Zr	15·2	0·152	109·5

* Cracking may not, of course, be proportional to these factors, but nevertheless they serve as a good guide to their order of harmfulness.

Class 'O'

Elements in this group are insoluble in solid iron. The liquidus and solidus are represented by horizontal lines extending from 0 to 100% solute.

Class 'OO'

These elements form typical monotectic systems.

Class I

All these elements form gamma loop systems. Solubilities are greater in α than γ iron but only α iron solidifies directly from the melt. The extent of the freezing ranges with type 1 elements varies considerably from short to very long, but with type 2 elements the freezing ranges are all very short.

Class II

These elements are more soluble in γ than in δ iron; both γ and δ iron can exist in equilibrium with a liquid phase. Type 1 elements form short delta and long gamma freezing ranges and type 2 elements short delta and also short gamma freezing ranges.

Class III

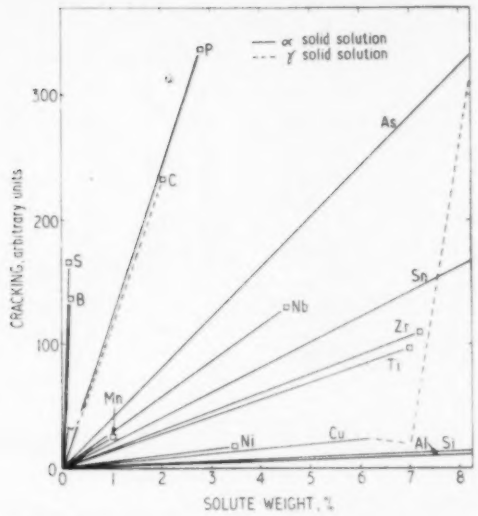
As the Class I group, elements here are more soluble in δ than γ but in this case only δ iron freezes from the melt.

Class IV

Here both γ and δ iron can exist in contact with liquid, but unlike Class II elements, solubilities are greater in δ than γ iron. An additional feature of importance is that for three of the four elements in this group the solubilities are very small.

Cracking tendencies

Assuming that the extent of the freezing range is roughly proportional to the amount of solute (up to the solid solubility limit) and to the severity of cracking, then it is possible to discuss the relative effects of various elements in terms of the R.P. factor as has been done for binary aluminium alloys (*see* Table II). The discussion is, however, necessarily hypothetical as no weld data exists. Figure 6 shows the idealised correlation of cracking and composition. Note the detrimental effects of sulphur, boron and phosphorus, and minor effects of nickel, silicon and aluminium when present by themselves.



6—*Effect of solute concentration on cracking, assuming that extent of freezing range is directly proportioned to cracking*

Class 'O'

Cracking troubles should not be experienced in these alloys as on initial solidification the liquid contains 100% solute. Presumably in many cases, an immiscible gap exists above 1534°C. Elements in this group include Li, Na, Mg, K, Ca, Sc?, Rb, Sr, Y?, Ag, Cd, In, Cs, Ba, La?, Hg, Tc, Pb, and Bi.

It is interesting to note that liquid Ag, Cd, Pb, and Bi form dihedral angles of near 180° ($\tau > > 5$). Values for the other elements do not appear to have been determined.

Class 'OO'

Oxygen is the only element in this group, and, as the maximum freezing range is only about 14°C. for oxygen contents below 0·16%*, cracking probably does not occur. In amounts greater than 0·16% the elements are immiscible above the monotectic temperature (1523°C.).

Class I

(i) *Type 1·1.* Phosphorus has the lowest solubility (2·8%) and is followed by Ti (7%), Be (7·4%), As (11·0%), Sn (16·1%), and Si (18·5%). Here phosphorus with its high R.P.F. (121) should be the most harmful, followed by arsenic (R.P.F.=40·2) and tin (R.P.F.=16·8).

Silicon (R.P.F.=1·75) should not be harmful.

(ii) *Type 1·2.* Aluminium, (R.P.F.=1·52) the only element here, should not be harmful.

(iii) *Type 2·1.* Molybdenum and tungsten have large solubilities. (37·5% and 33·0% respectively) and very narrow freezing ranges and thus should not cause cracking.

(iv) *Type 2·2.* Both elements (vanadium and chromium) in this group form continuous solid solutions with iron and as type 2·1 elements have narrow freezing ranges; no cracking should occur.

* Solubilities in weight %.

Class II

(i) *Type 1·1.* Carbon, the only element of this type, has a low solubility in both δ (0·10%) and γ (2·03%) iron but in amounts up to about 0·16% should not be very harmful because of the small delta freezing range; above this value the extended gamma freezing range (max.=200°C) will assume more importance but carbon will probably not be too harmful until it is present to about 0·51% when only γ iron solidifies from the melt.

(ii) *Type 1·2.* Of the two elements here only copper is important. Gold, for a good reason, is not used as an alloying element. Copper has a relatively large solubility in δ (6·25%) and γ (8·5%) iron, and should not be harmful until present in amounts approaching 8·5% when the gamma phase with its long freezing range is formed.

(iii) *Type 2.* Only manganese (type 2·1) and nickel (type 2·2) are used as alloying elements; both have a relatively low solubility in δ (1·0% and 3·4% respectively) iron. As type 1 elements the freezing ranges are small and cracking should not occur; the gamma freezing ranges are even smaller and continuous solid solutions are formed.

Class III

Delta is the only phase which forms directly from the melt. Niobium with a R.P.F. of 28·8 is more harmful than tin but less harmful than arsenic. Tantalum (R.P.F.=13) should be almost similar to tin.

Class IV

Two elements of this type, namely boron (R.P.F.=917), and sulphur (R.P.F.=927) have very small solubilities in both δ and γ iron and their attendant freezing ranges are large. Their solubilities in δ iron are 0·15% and 0·18% and in γ iron 0·021% and 0·013% respectively. Of all the elements these should be the most harmful.

Cerium also belongs to this group but as the binary diagram is not known accurately, the effect of this element can not be properly judged. The solubility in γ iron is about 0·4% at 1060°C.

The fourth element, zirconium, has a R.P.F. of 15·2 and thus should be as harmful as tin. The maximum potency of tin (270) is however twice that of zirconium.

Summarising elements which should be very harmful because of their long delta and gamma freezing ranges even when present in very small amounts are sulphur and boron and perhaps cerium. Phosphorus should be the next most dangerous and is followed by arsenic, niobium, tin, zirconium and tantalum. Elements which are harmful only if present in sufficient amounts to cause gamma iron to exist in contact with a liquid phase, are carbon and copper.

Cracking Tendencies of Elements in Mild and Low Alloy Steel

Significance of the dihedral angle

In binary solid solution alloys where the dihedral angle is only likely to attain values of just above 0° it has been possible to relate the severity of cracking

simply to the extent of the freezing range. With the more complex alloys, where large differences between the solid phase and grain boundary liquid can develop because of the ability of certain elements to combine chemically with, dissolve, or be soluble in, other elements or phases, the dihedral angle will assume more importance. Here the relationship between cracking and the extent of the freezing range is much more complex.

Studies[3–5] on non-weld samples brought to near equilibrium, have shown that, generally, when a third element is added to a binary alloy, the dihedral angle will increase if the third element is more soluble in the liquid than the solid. (Conversely, increasing mutual solubility of the solid and liquid phases lowers the interfacial energies.) Whether this occurs under welding conditions is a matter for experiment. The fact that high dihedral angle phases are sometimes found in weld samples would indicate that it does in certain cases, notably when FeO is present. The chances of this happening are increased presumably, when there is a definite chemical interaction (*e.g.*, combination of Mn and S to form MnS), and also when the events take place just below the liquidus. Measurements[4,5] of θ, under equilibrium conditions, in a pure iron–sulphur alloy (0·12%S) have been made and it was found that at 988° θ was 30°; it decreased to 20° at 1200°C. and to zero at 1300°C. Adding manganese increased θ and also the liquid melting temperature. Silicon (1%) appeared to have little or no effect on the sulphide distribution. When aluminium was added the sulphide did not form a continuous film at 1300°C. as did the aluminium-free samples; instead a chain type structure was formed with individual sulphide particles having dihedral angles of 20–30°. With small amounts of oxygen the dihedral angle at 1300° was just finite thus indicating that oxygen only slightly modified the liquid iron interfacial energy.

As during welding equilibrium conditions are not obtained, the forgoing can only serve as a guide and these events must be interpreted remembering that segregation occurs during the solidification process and dictates the amount of solute which enters the grain boundary liquid. Certainly elements forming low dihedral angles should be harmful in welds.

Cracking in mild steel

Mild steel is essentially a low carbon iron containing small amounts of manganese and silicon. Sulphur and phosphorus are also present in amounts up to 0·05% each and are regarded as harmful impurities as they are known to cause cracking. Manganese is beneficial and reduces the effects due to sulphur. Manganese also acts as a deoxidiser (weak) as does silicon (strong). Carbon, the essential ingredient, is known to promote cracking when present in amounts exceeding about 0·15% wt.

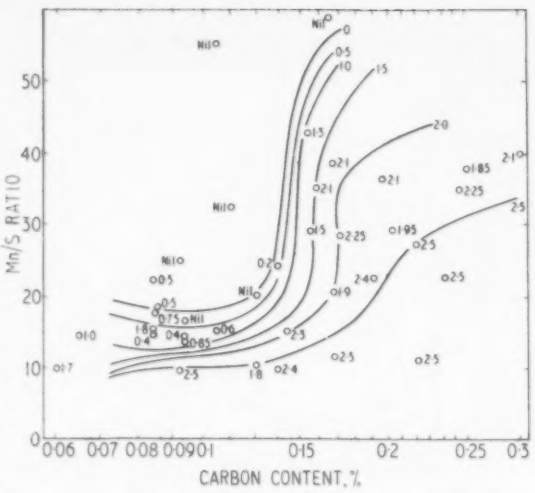
Since the general effects of carbon, manganese and sulphur are widely known[11] and to a large extent interrelated it is proposed to discuss these elements first.

All the weld data discussed refer to the metal arc welding process unless otherwise stated.

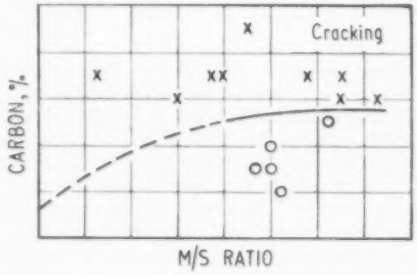
Effect of carbon, manganese and sulphur

The interrelated effect of carbon, manganese and sulphur on cracking in mild steel weld metal has been investigated by Podgaetskii[12] and Jones[13] (Figs. 7 and 8). Podgaetskii using a single Tee weld cracking test found that all welds containing more than 0·16%C cracked irrespective of the Mn/S ratio (maximum ratio investigated was 40). Jones, using the Murex hot cracking machine, investigated the effect of the Mn/S ratio up to a value of 59. Jones concluded that "in the range 0·06–0·11%C, the Mn/S ratio is the controlling factor, crack-free welds being obtained for Mn/S ratios exceeding about 22. Above 0·11%C carbon content becomes more important but freedom from cracking is still obtainable by increasing the Mn/S ratio to 30. With carbon contents over 0·13% cracking is more prevalent, but high Mn/S is still beneficial even at the highest level of carbon investigated. The highest carbon content for which a completely crack-free weld was obtained was 0·155% with a Mn/S ratio of 59."

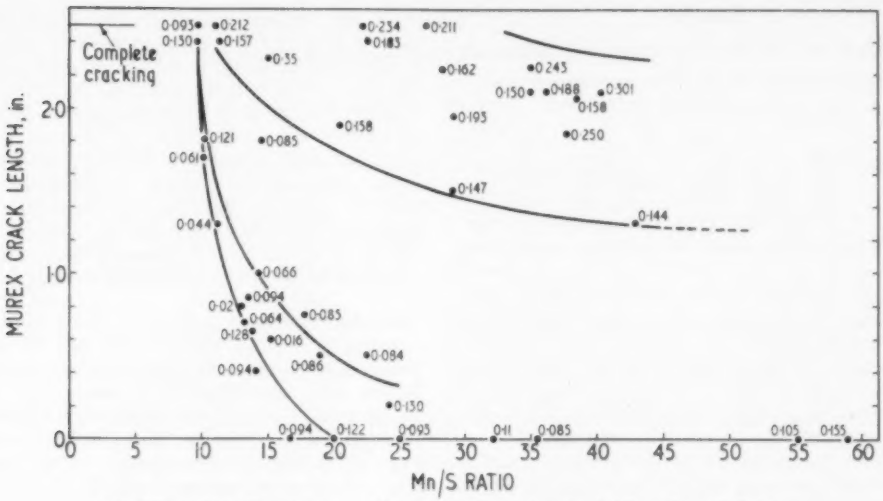
An alternate method of plotting Jones' results (given in his Table I) is shown in Fig. 9 and demonstrates that (*a*) at a carbon content of about 0·08–0·09% cracking is apparently more severe than at a higher carbon level (up to 0·13%); (*b*) increasing the carbon content from 0·13 to 0·15% produces a marked change



7—*Effect of carbon content and* Mn/S *ratio on severity of hot cracking in Murex test on mild steel plates*[13]



8—*Influence of carbon, sulphur, and manganese contents of weld metal on hot cracking tendency during submerged-arc welding, using low silicon flux, free from manganese*[12]



9—*Effect of* Mn/S *ratio on cracking for various carbon levels (Data from Ref. 13)*

in the cracking susceptibility; and (*c*) at and above 0·15% carbon increasing the Mn/S ratio only reduces cracking slightly.

(Note: The nil crack length obtained by Jones at 0·155% carbon with a Mn/S ratio of 59 warrants further investigation).

These results are interpreted by the present author as follows:

0·084–0·13% *carbon*

At low Mn/S ratios, sulphur is the element primarily responsible for cracking and annulls any effect due to carbon. At high Mn/S ratios there is a small effect produced by carbon and it is associated with a widening of the delta freezing range which probably reaches a maximum at about 0·07–0·085% carbon. Above this level the shortening of freezing range causes a reduction in cracking.

Above 0·13% carbon

Above 0·13% carbon there is an extremely rapid increase in cracking, which appears to be associated with the peritectic composition (0·16% carbon), and although the extent of the freezing range starts to increase again at this point, this cannot be the reason for the marked increase in cracking (rapid rather than gradual), as the actual increase is very much greater than is to be expected from consideration of freezing range concepts alone.

There is no doubt that carbon is instrumental in causing cracking and since it is a strong austeniser (class II element) it seems likely that *its chief role is to increase the amount of* γ *iron initially solidifying, and to raise the* $\gamma/\delta+\gamma$ *phase boundary thus bringing the liquid into contact with* γ *iron at a higher temperature.* Also because of the increased solubility of manganese and reduced solubility of sulphur in γ iron, the net effect of adding carbon is to increase the relative amounts of iron and sulphur in the grain boundary liquid thus decreasing θ and increasing the freezing range. This will increase the severity of cracking.

The dramatic increase in cracking for only a small addition of solute is similar to the situation experienced with chromium–nickel austenitic steels where, as the last amounts of delta ferrite vanish (as observed at room temperature) the crack sensitivity increases rapidly.

At carbon contents of greater than about 0·15% where high Mn/S ratios have only a small effect it is reasonable to suppose that other elements which are not counteracted by manganese (*e.g.*, phosphorus) are exerting a much greater effect because of their reduced solubility in austenite. Carbon itself will, of course, cause cracking if present in sufficient amounts to widen the freezing range. On the basis of its effect in delta iron, it should be possible to detect a small effect due to carbon at about 0·3%.

Manganese, being a class II element, is also an austeniser, but in comparison with carbon it is much less potent. Furthermore the manganese is to some extent rendered less effective by combining with sulphur, and in some cases oxygen. The relative austenising powers of carbon and manganese may be estimated by assessing their ability to form austenite. Using the peritectic composition as a basis for comparison then the relative austenising powers are:

$$C=1 \quad Mn=\frac{1}{38}$$

Phosphorus

There is some evidence[11] that phosphorus increases hot cracking in mild steel weld metal; detailed data are lacking however and do not permit a critical appraisal to be made. Rollason and Roberts[14] using the Murex hot cracking machine found that the crack length increased from zero to 1·7 in. when the phosphorus was increased from 0·039% to 0·072% (Table III). It is to be noted that the carbon content of the authors' samples was 0·125–0·130% which indicates that the Mn/S ratio (greater than 23) should have been sufficient to prevent sulphur causing cracking. However in view of the nearness of the carbon content to the critical 0·13% level, this result seems worthy of a

Table III

Effect of silicon and phosphorus on Murex hot crack length[14]

Effect of:	*Code*	*Weld Metal Analysis, % wt.*					*Crack Length in.*
		C	*Mn*	*S*	*P*	*Si*	
Phosphorus	S	0·13	0·55	0·02	0·039	0·15	0
	F	0·125	0·51	0·022	0·072	0·15	1·7
Silicon	Y	0·11	0·50	0·026	0·056	0·135	0
	R	0·10	0·53	0·128	0·042	0·162	2

further study. Stieler[15] also found that 0·071% phosphorus caused cracking when making welds, but in his case the weld metal carbon content was 0·14% and the Mn/S ratio 16·5. It is likely, therefore, that sulphur was a contributory factor causing cracking. Reeve[16] did not encounter cracking in welds made in restrained joints with phosphorus contents up to 0·128%.

It is interesting to note that the silicon content of Reeve's samples were all much less than 0·1% suggesting that FeO inclusion existed in the weld (in the case of the 0·128% phosphorus the silicon content was 0·033%).

Helin and Svantesson[17] comment that in amounts up to 0·07%, phosphorus has only a slight effect in normal weld metals. No data are offered.

According to the assumptions already made in this paper, the R.P.F. for phosphorus in the binary system is 121 and is 7·4 times smaller than that of sulphur, the next most harmful element. Therefore in a low carbon steel (<0·13%C) and in the absence of the complicating effects of manganese, sulphur and silicon, phosphorus would not appear to be very harmful – certainly not as harmful as sulphur. In a high carbon steel (>0·13%C) when austenite is one of the 6 phases initially solidifying, phosphorus would be more potent, but as yet it is not possible to say how much.

Arsenic

Using the Reeve fillet weld, and controlled thermal severity tests, one author[18] found that cracking occurred when the arsenic content of the weld bead exceeded about 0·2%. With the Murex hot crack machine he found a detectable effect at 0·17%. This level of arsenic is of course, 4–5 times higher than is present in normal steels, and as such would not be harmful in normal welding practice.

Arsenic would tend to act like phosphorus in its manner to cause cracking, though in view of its lesser tendency to segregate (R.P.F.=40), it should be less harmful.

Silicon

Rollason and Roberts[14] suggest that silicon is harmful because of its tendency to increase the segregation of sulphur. These authors, however, do not present firm evidence to support their statement. In fact the data they gained to show the effect of silicon do not appear to be valid as their higher silicon weld metal also contained 0·128% sulphur, giving a Mn/S ratio of only 4·5 (carbon=0·10%) (Table III).

Reeve[16] whilst thinking that silicon assisted sulphur in segregating also suggested that silicon was harmful owing to its deoxidising action. This aspect will be discussed in the next section on the effect of oxygen.

In the present author's view, silicon by itself in amounts up to 1·0% in a low carbon (<0·13%) steel, is only slightly harmful, for the maximum extent of the freezing range only amounts to about 50°. It is also slightly beneficial as it increases the solid solubility of carbon (about 0·15%C at 1·0%Si), and being a class I element also lowers the $\delta+\gamma/\gamma$ phase boundary. This latter beneficial effect would however only be noticed when an element other than silicon (*e.g.*, S) was harmful. These benefits are obviously not conferred by other class I elements such as phosphorus and arsenic, as they are harmful even in ferritic steels.

Oxygen

Oxygen in mild steel should not be harmful, as below the monotectic temperature of 1523°C. the liquid contains about 22% oxygen, and is very different from the iron solid solution which contains much less than 0·01% oxygen and the iron oxide formed tends to ball up as a consequence of the high dihedral angle developed. Iron oxide (FeO) globules are good examples of this.

Iron oxide having a certain solubility for iron sulphide should tend to reduce cracking caused by sulphur, as in fact the work of Swinden and Tremlett[19] and Reeve[16] shows. The paper by Swinden and Tremlett on the welding of high sulphur free cutting steels (0·24%S) is particularly interesting as in preliminary experiments they noted that a highly oxidised, but manganese free weld deposited under an oxidising slag was better as regards crack resistance than others containing manganese deposited under normal slags. Following this lead, they investigated several types of dead soft and mild steel electrodes and found that all the welds made with the mild steel electrodes contained visual cracks, whereas welds made with the dead soft electrodes were free from such cracks. Reeve in a later investigation, also found that dead soft (oxidised) weld metal was extremely resistant to the effects of normal amounts of sulphur. Increasing the silicon content was harmful due to the higher state of deoxidation. Examining his welds Reeve found that the uncracked samples contained greater quantities of sulphur combined with the larger oxide globules, and he suggested that this prevented intergranular networks of FeS which were regarded as the cause of cracking.

It is for these events that the 'generalised theory' is most applicable and explains why oxygen can be beneficial and silicon detrimental. Oxygen of course, is not harmful because the dihedral angle increases to a large angle during the first moments of solidification. Further, it is beneficial when it is able to dissolve an element which is normally harmful (*e.g.*, S). If however it interacts with a third element (*e.g.*, Si) to form a phase or compound which has a much reduced solubility for the harmful element, it is rendered ineffective. Manganese is also a deoxidiser and as it reduces the amount of FeO inclusions in weld metal[20] it should act in a similar manner to silicon. In doing this however it is less potent than silicon, for the following reasons: (*a*) it is a weaker deoxidiser, and (*b*) it tends to form (MnFe)S which has a higher melting point and larger dihedral angle than FeS. Thus the first additions of manganese increase cracking whilst further amounts reduce cracking. Silicon

Table IV

Effect of chemical composition on crack resistance (Single-Tee Test)[12]

Weld Metal Analysis, wt.%									
C	*Si*	*Mn*	*S*	*P*	*O*	*N*	*Mn/S Ratio*	*Cracks*	*(C)×(O)*
0·13	0·11	0·81	0·031	0·025	0·0316	0·0050	26·2	No	0·0041
0·15	0·07	0·48	0·035	0·028	0·0291	0·0051	13·7	,,	0·0044
0·16	0·18	0·87	0·034	0·027	0·0366	0·0038	25·5	,,	0·0058
0·19	0·43	0·88	0·040	0·030	0·0300	0·0065	22	Yes	0·0057
0·30	0·19	1·12	0·057	0·054	0·0273	0·0051	18·3	,,	0·0082
0·33	0·44	1·29	0·027	0·030	0·0178	0·0026	48	,,	0·0058

does not reduce cracking because it does not combine with sulphur.

It has been thought that the effect of carbon on hot cracking is also related to its deoxidising action. However, Podgaetskii[12] carried out some experiments to investigate this and concluded that cracked welds contained less oxygen than those without cracks but the difference in oxygen content between the two groups was very small (Table IV). He does not, however, comment on the nature of oxides formed. Presumably his figures indicate 'total' oxygen. Podgaetskii concluded that carbon itself was exerting a much greater effect than oxygen. Since the silicon content of the welds varied considerably, the data for the samples that did not crack are worthy of further study. For instance at a Mn/S ratio of approximately 26 the welds containing 0·13% and 0·16% carbon did not crack. Normally the higher carbon weld should be crack sensitive. (See Podgaetskii line—Figure 7). The silicon content of this weld was 0·18%; the other uncracked weld contained 0·11% Si. Another weld containing 0·15% carbon and 0·07% silicon also did not crack despite its relatively low Mn/S ratio (13·7). Perhaps this weld contained iron oxide (with sulphides) in the globular form.

It is quite evident that it is not the oxygen content itself which is important but the quantity, manner and distribution in which the oxides exist. Deoxidation practice in steelmaking might therefore play an important role in hot cracking. This aspect is very worthy of a detailed study.

Aluminium

Aluminium is a very strong deoxidiser and is often added to steels as a grain refiner. In view of its deoxidation characteristics, it should, therefore, act in a similar but more potent manner to silicon. In steel castings there is some evidence[21] for this. For instance when present in amounts exceeding about 0·005% aluminium is very harmful because it produces Type II (intergranular) sulphide films; greater amounts of aluminium produce the less harmful Type III (angular) particles. Presumably the first additions of aluminium deoxidise the metal, thus removing the beneficial effects of oxygen, and forming an 'active' FeS of low dihedral angle whilst further additions tend to combine with sulphur forming (AlFe)S of higher dihedral angle. The melting point of AlS is 1200°C. and is not much higher than that of FeS (1190°C.).

Few data are available for the effect of aluminium on hot cracking in mild steel welds. Jones,[13] using the Murex hot cracking machine concluded that "the effect of aluminium on hot cracking does not appear to be very great" (Table V). Gregory,[22] on the other hand, in an examination of some welded rimming steel components, found that aluminium caused the formation of inter-granular films and consequent cracking when present in amounts somewhere in the range 0·0075–0·040%. The Mn/S ratio was about 11–13, and on the basis of Podgaetskii[12] results would not be sufficient to stop cracking in a *deoxidised* (silicon) weld. The Mn/S ratio of the welds examined by Jones may have been much greater. Alternatively his aluminium content may have been too high to show a detrimental effect.

Table V

Preliminary Murex hot-cracking tests on mild steel plates with aluminium additions[13]

BS.1719 Class	*Electrode Code*	*Plate*	*Al in Weld Metal, %*	*Crack Length, in.*
E219	B	Calorised mild steel	0·03	0·8
E219	B			0·8
E319	E		0·07	1·55
E319	E			1·3
E616	AO		0·14	nil
E616	AO			nil
E219	B	Mild steel		0·3
E319	E			1·75
E616	AO			nil
E616	AO			0·2

Cracking in low alloy steels

For the purposes of this paper it is considered that a low alloy steel is one which contains less than about 6·0% total alloy additions. The addition of elements in these amounts is assumed not to affect the solubility of other elements. This would not be the case in high alloy steel.

Nickel

Rollason and Roberts[14] and Jones[13] using the Murex hot cracking machine found that nickel additions to mild steel caused increased cracking. (Tables VI and VII). According to Jones the electrodes used varied considerably in their tolerance for nickel and although there was a trend for weld metals of high Mn/S ratio to withstand more nickel without cracking than those with a low Mn/S ratio, the correlation was poor.

A completely satisfactory explanation for the effect of nickel has not been advanced. Rollason and Roberts suggest that for a given sulphur content, nickel is instrumental in increasing the amount of grain boundary sulphide and consequent cracking. Williams *et*

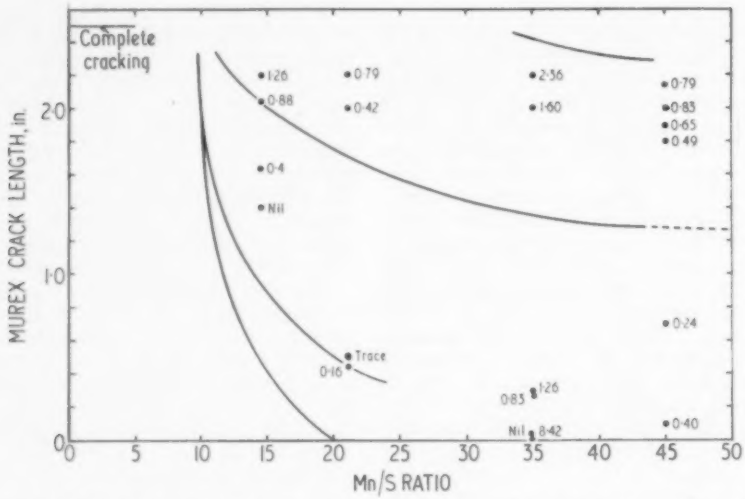
Table VI

Effect of nickel on Murex hot crack length[14]

Cracking Length (Ni nil), in.	*Wire Ni Content, %*	*Cracking Length, with Ni, in.*
½	E.317 0·08	½
½	E.317 0·32	1¾
1	E.317 0·60	3
	Coating Addition Weld Ni %	
Low Mn (0·35) deposit 1 in.	0·5	3
High Mn (0·80) deposit 0 in.	1·5	1½

al[11] suggest that its effect may be related to the low melting point of NiS (920°C.).

The results of Jones have been plotted (by the present author) on the diagram showing the effect of carbon and the Mn/S ratio on cracking (Fig. 10); it will be seen that the results for welds containing more than about 0·5% nickel are generally similar to those containing 0·15% or more carbon. Also the crack length increases rapidly at the appropriate nickel content rather than gradually as would happen if nickel were merely increasing the freezing range. Thus it seems that nickel is acting as an austeniser like



10—*Effect of* Mn/S *ratio on cracking for various nickel contents* (*See Table VII*) (*Data from Ref. 13*)

Table VII

Murex hot-cracking tests on mild steel plates with nickel wire additions[13]

*Carbon** %	*Electrode Class*	*Code*	*Ni Wires Added No.*	*Dia., in.*	*Composition, % Mn*	*S*	*Ni*	*Crack Length, in.*
0·061–0·093	E.319	E	None	—	0·42	0·029	nil	1·4
			1	0·0164			0·40	1·65
			2	0·0164			0·88	2·05
			3	0·0164	0·39	0·027	1·26	2·2
0·094	E.646	AP	none	—	0·49	0·024	trace	0·5
			1	0·0116			0·16	0·45
			1	0·0164			0·42	2·0
			2	0·0164	0·51	0·024	0·79	2·2
0·066–0·084	E.614	N	none	—	0·62	0·020	nil	nil
			1	0·0164			0·42	nil
			2	0·0164			0·83	0·3†
			3	0·0164			1·26	0·3†
			4	0·0164	0·70	0·018	1·60	2·0
			6	0·0164			2·36	2·2
0·093–0·111	E.616	AO	none	—				nil
			1	0·0116			0·24	0·7
			1	0·0164			0·47	1·8§
			2	0·0164			0·65	1·9
			3	0·0164	0·86	0·019	0·79	2·15
0·105	E.616	AQ	none	—	1·02	0·024	trace	nil
			1	0·0164			0·40	0·1
			2	0·0164			0·63	nil
			3	0·0164	0·95	0·020	0·83	2·0

* Not in original table. † Porous start § Porous weld

Table VIII

Chemical composition of commercial steel plate investigated by Jones[13]

Designation	*Type*	*Composition, %*								
		C	*Si*	*S*	*P*	*Mn*	*Ni*	*Cr*	*Mo*	*B*
Fortiweld II	Mo	0·11	0·19	0·040	0·034	0·55	—	—	0·49	0·0028
D	Mn–Ni–Cr–Mo	0·16	0·22	0·039	0·016	1·05	1·13	1·02	0·31	
Ducol W 25	Mn–Mo	0·19	0·05	0·025	0·019	1·48	0·33	0·10	0·28	
E27(HC)	Ni–Cr–Mo–V	0·51	0·23	—	0·014	0·55	3·30	0·71	0·65	

carbon. The relative power of nickel is 1/28 and so 0·5%Ni is equivalent to 0·018%C. Whether the effect of nickel is additive to carbon is not known for certain. However the range of carbon contents of the welds investigated by Jones is included in Table VII for comparison, and it will be seen that the electrode having the best tolerance for nickel could have had the lowest carbon content. Two tests on a steel plate (designation 1D) containing 2·34%Ni and 0·21%C gave crack lengths of 0·3 in. and nil, thus possibly indicating that in amounts up to about 1% (assuming 50% dilution) nickel had no effect. Unfortunately, no weld compositions are given.

Another effect of nickel may be to replace Mn in the (FeMn)S lowering the melting temperature and dihedral angle.* The sulphides found in mild steel welds containing 0·60%Ni examined by Rollason and Roberts[14] resisted balling up at 1000°C.

Copper

The American Welding Handbook states that "weld cracking difficulties have been attributed to copper but this depends on the process used and the remaining composition of the steel". Adding 0·99% copper to mild steel Jones[13] found that the cracking susceptibility increased slightly.

According to the generalised theory copper (1%) should not be harmful in the iron binary system which initially solidifies as δ iron.

In mild steel however the effect is probably very much dependant on the amount of austeniser present, particularly carbon, as these elements raise the austenite/ferrite phase boundary, thus bringing copper containing liquid into contact with austenite at higher temperatures.

It is interesting to note at this juncture that the interfacial energy of liquid copper sulphide (Cu_2S) against gamma iron (1105°C.) is 470 ergs/cm² as opposed to that of 430 ergs/cm² for liquid copper. This means that Cu_2S sulphide forms a higher dihedral angle than does copper or sulphur. In certain circumstances therefore it seems possible that copper and sulphur by themselves might be bad whereas when they are present together in the required proportions, they would not be harmful.

The austenising power of copper relative to carbon is approximately 1/50.

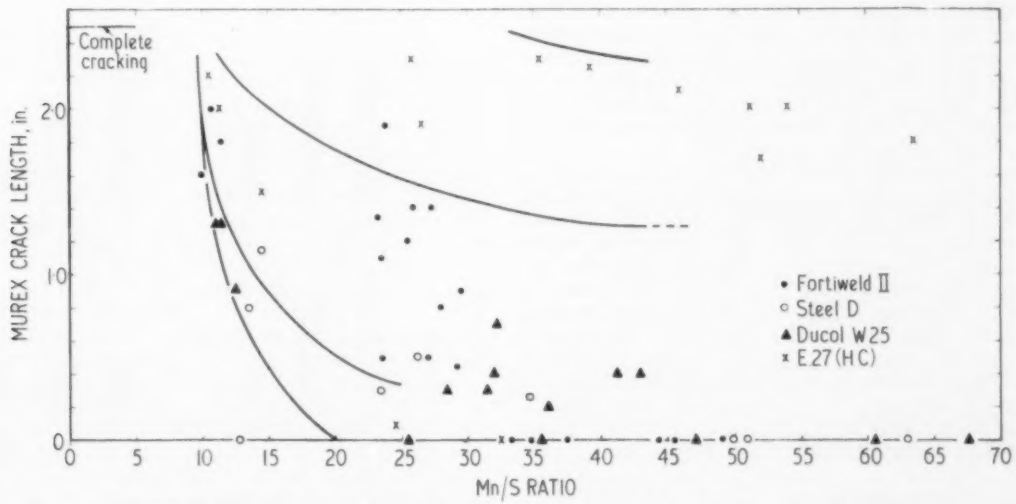
Chromium

Ad hoc experiments by Jones[13] indicate that 0·83% chromium does not increase cracking.

Molybdenum

According to Apold[23] molybdenum often helps to counteract crack sensitivity and is due to its effect in

* The effect of nickel in increasing the dihedral angle is presumed to be less potent than manganese.



11—*Effect of* Mn/S *ratio on cracking in commercial steels* (*See Table VIII*) (*Data from Ref. 13*)

Table IX

Chemical composition of experimental steels[25]

Heat	Composition, % C	Mn	Si	Ni	Cr	Mo	S	P	N	Al
A	0·20	0·67	0·30	2·07	0·91	0·38	0·008	0·010		
B	0·40	0·70	0·33	1·91	0·78	0·29	0·004	0·004		
C	0·59	0·66	0·29	1·93	0·81	0·26	0·006	0·007		
D	0·40	0·76	0·22	1·84	0·83	0·24	0·015	0·012		
E	0·35	0·77	0·23	1·84	0·81	0·23	0·036	0·011		
F	0·43	0·86	0·33	1·86	0·86	0·26	0·006	0·017	0·005	
G	0·38	0·84	0·32	1·85	0·88	0·26	0·007	0·039	0·006	
H	0·40	0·58	0·25	1·71	0·80	0·24	0·031	0·009	0·006	0·03
I*	Split heat						0·025			
J†							0·014			

* 0·15% rare-earth metal added to Heat H.
† 0·30% rare-earth metal added to Heat H.

conferring strength and toughness at high temperatures. No evidence is presented.

Commercial Steels and Experimental Alloys

Jones[13] has investigated the effect of the Mn/S ratio in commercial steels (Table VIII). His results, plotted in Fig. 11, show that steels *D* and Ducol *W*25 are similar to mild steel, containing less than 0·13%C; the boron-containing Fortiweld II is slightly worse, indicating that boron, even in very small amounts, is a potent crack former, and the high carbon (0·51%) and nickel (3·30%) steel *E*27 (*HC*) is similar to mild steel, solidifying initially mainly as austenite. Presumably, except for Fortiweld II, sulphur and phosphorus are mainly responsible for cracking, and the position of the steels on the diagram is determined by whether they solidify as ferrite or austenite. There is no definite 'in-between' position. Fortiweld II results, however fall in the intermediate position, and demonstrate that boron is roughly equally harmful in ferrite as it is in austenite. Since boron is neither a strong ferrite nor austenite former and is only slightly soluble in both this result is not unexpected.

In investigating* some experimental alloy simulating type SAE 43XX and SAE 4340 steels (Table IX) initially solidifying as austenite, Sopher *et al*[24,25] found that carbon (0·20–0·59%) had a minor effect, but sulphur and phosphorus were quite harmful in amounts greater than about 0·01% (Fig. 12). If these results are analysed it will be noted that sulphur and phosphorus are about equal in their effects, and that they start to become harmful when their combined total exceeds about 0·013%. Looked at in another way sulphur appears to decrease the crack resistance when the Mn/S ratio drops below about 60–70. This is interesting, as the plot of Jones' results for carbon contents in excess of 0·15% (Fig. 9) shows that the Mn/S ratio confers benefits up to about 50; thereafter it does very little in the way of reducing cracking. Sopher[25] also finds that additions of misch metal increase the cracking resistance, and that its action is due in part to the balling up of the remaining sulphides (Fig. 12). This is a further indication of the dihedral angle effect. Work by Wilkinson *et al*[26,27] also shows that sulphur and phosphorus are extremely harmful.†

* Consumable electrode process.
† Argonarc process.

Lowe *et al*[28] found that silicon (1·67%) when added to SAE 4340 steel was not harmful except in regard to cold cracking.

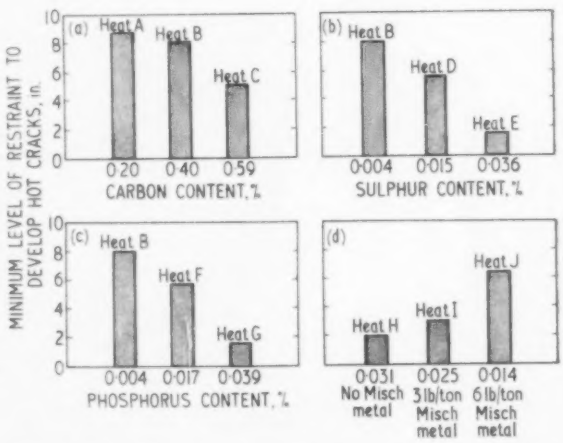
Discussion

To develop a general explanation of cracking it is useful to consider first the effects of elements in binary alloys particularly in their action in widening the freezing range. It is assumed that the severity of cracking is approximately proportional to the extent of the freezing range, which in turn is proportional to the C.S.R. where cracking occurs.

The various binary diagrams are divided into classes according to their constitutional features (Fig. 5) and the unit increase in the freezing range per unit weight solute $\left(M_L\frac{1-K}{K}\right)$ which is called the Relative Potency Factor has been calculated for several elements (Table II). These factors can also serve as a guide to the effect of elements in mild and low alloy steels.

Binary alloys

Elements in class 'O' system are not harmful as on initial solidification the liquid composition is 100%



12—*Effect of C, S, and P on cracking as measured by the Lehigh Restraint Test*[26,25]

solute, and dihedral angles of near 180° are developed. In class 'OO', of which oxygen is apparently the only member, cracking should not normally occur even for alloys containing less than 0·16% oxygen because at worst the harmful condition (θ near 0°) only exists for about 14°C. Further, below the monotectic temperature the dihedral angle should increase to a large value. If however, because of the rapid rate of cooling the 'almost continuous' film state persists below the monotectic temperature, cracking could probably be very severe. At oxygen contents above 0·16%, the immiscibility of the components should prevent films being formed. In class I systems the most harmful elements are phosphorus and arsenic which have R.P. Factors of 121 and 40 respectively. Next comes tin (R.P.F.=16) and then titanium (R.P.F.=14). Silicon (R.P.F.=2) and aluminium (R.P.F.=1·5) are only very slightly harmful and under normal welding conditions should not cause cracking. Elements in class II alloy are characterised by their short delta freezing ranges and except perhaps for carbon which has a maximum freezing range of about 32, should not cause cracking when present in amounts not exceeding the peritectic composition. Above this, type 1 elements, carbon and copper form long gamma freezing ranges and are thus harmful, particularly copper which has an almost vertical solidus at 8·5%Cu. Type 2 elements (Mn, Co, Ni, Pd, and Pt) have very short gamma freezing ranges and should not cause trouble. Niobium and tantalum, the only elements in class III systems, have R.P. Factors of 29 and 13 and should be a little less harmful than arsenic. In class IV sulphur and boron have extremely high R.P. Factors, being 925* and 917 respectively. Zirconium has a factor of 15. The value of cerium is not known. However on the assumption that cerium belongs to this group, the maximum solid solubility in delta ferrite must be greater than 0·4% and the R.P. Factor not more than 354.

Summarising, the most harmful elements in decreasing order of potency are:

S B (Ce?) P As Nb Sn Zr and Ta

(Note. No class II elements are included).

Mild and low alloy steels

The elements which can cause cracking in binary alloys will also be, probably, as harmful *per se* in alloys initially solidifying mainly as ferrite. Now, as their solubilities are less in austenite than ferrite, they will be more harmful in alloys mainly solidifying as austenite, though the order of potency might be slightly changed. In addition other class I type 1 elements such as silicon might also assume more importance though to the author's mind they will be very much less potent than phosphorus. Carbon and copper, the type 1 elements in class II systems could also be harmful, particularly copper.

Of all the elements normally occurring in commercial steels sulphur and phosphorus are probably the most prominent and are the elements mainly responsible for cracking.

Class II elements can act as crack promoters as they are austenite formers and tend to replace the less crack-sensitive ferrite structures. However, of these elements carbon and nickel are the only ones powerful enough in small quantities to produce a significant effect. By the same reasoning class I elements being ferritisers should be beneficial when added to austenitic alloys. Here, molybdenum, tungsten, vanadium and chromium should confer the most beneficial effects. On the other hand class I elements like phosphorus and arsenic should not be added as they themselves are harmful in ferritic structures. This situation is similar to that experienced in austenitic stainless steels where some ferrite forming elements are harmful unless in sufficient quantities to form ferrite.[29]

Certain elements can act to promote or reduce cracking in other ways. For instance oxygen is beneficial because of the large dihedral angle developed and its ability to absorb a certain quantity of FeS. Additions of aluminium and silicon are detrimental as they are deoxidisers and form oxides thus reactivating the FeS. Excess quantities of aluminium should tend to reduce cracking as it combines with sulphur. Silicon does not. Other sulphide formers which are beneficial are La, Ce, Zr, Ti, and Mn. Here La should be the most useful element because of its high heat of formation (−125,000 cal/g.mol) and high melting point sulphide (>2000°C.) which ensures that it is precipitated early in the solidification process.† Furthermore being a class 'O' element, lanthanum in excess quantities should not be harmful. Cerium, another element which tends to form a high melting point sulphide and has a high heat of formation (−114000 cal/g/mal) should also be very useful. However in view of it being a class IV element it could be harmful in excess amounts. It might be thought also that zirconium and titanium should be better than manganese additions since they produce more stable sulphides than manganese. However since it segregates more during solidification, manganese is likely to be the more beneficial.

It also seems likely that effect of manganese is not always associated with MnS which has a melting point of 1610°C. but in many cases its main effect will be to form a mixed sulphide (FeMn)S which has a higher melting point and dihedral angle than FeS, both effects contributing towards the reduction in cracking obtained. The melting point of NiS (920°C.) is lower than that of FeS and presumably would form a low melting sulphide of high dihedral angle, however adding nickel, stabilises the austenitic structure which makes matters worse.

Preventing cracking in mild steel

Jones[13] has suggested a weld metal composition which should not be prone to cracking. This is:

C	=	0·15 max.
Mn	=	0·80 min.
S	=	0·035 max.
Mn/S ratio		35 min.
Ni	=	1·0 max.

(Note that with 0·80 Mn and 0·035 the Mn/S ratio is about 23).

Re-interpreting the results gained by Jones the following composition is suggested:

* In very dilute solution this value can be 2–3 times higher.

† It is rather interesting also to note that lanthanum has a very high solubility for hydrogen being about 2000 times that of hydrogen at 1000°C.

C = 0·12 max.
Mn = 0·50 min.
S = 0·035 max.
P = 0·040 max.
Mn/S ratio 14·3 min.

This should be suitable for welds made under low restraint. For more severe conditions, and where good impact properties are desired the manganese should be increased, 0·8–1% being sufficient in the majority of cases.

Preventing cracking in low alloy steels

Ideally, the most suitable weld metal composition would be one initially solidifying mainly as ferrite. The ratio of austenisers and ferritisers should thus be balanced to obtain this. Where this is impossible, then the sulphur and phosphorus should be reduced to a very low level; in some applications where high strength and good notch toughness is desired the amounts should not exceed 0·005% each, the Mn/S ratio being greater than about 70. Some additional benefit, particularly in regard to weld metal properties, may be obtained possibly by also reducing the arsenic and antimony content, though experimental evidence for this has not yet been obtained as far as the author is aware.

It is not possible to recommend weld metal compositions which should be free from cracking as suitable data do not yet exist. However it is obvious that high carbon and nickel* contents are harmful unless sulphur and phosphorus contents are strictly controlled. Molybdenum, vanadium and chromium should be beneficial, as should also be manganese, lanthanum, and cerium in controlled amounts. Boron should be harmful.

* It would be interesting to know the solubility of carbon in an alloy containing 3–4%Ni. The effect of carbon at this level may not be additive, in fact the solubility of carbon might be relatively high.

Finally, it is generally recognised that the effect of minor elements becomes more important as the strength of the material is increased; the steels discussed here should not be an exception to this rule. Thus in ultra-high strength steels it is likely that the permissible limits of sulphur and phosphorus will decrease; also it is likely that elements such as arsenic and tin will assume more importance.

Normally class 'O' elements should not be harmful; it is a possibility however that in certain alloys containing perhaps a high proportion of a major element, these insoluble elements will become slightly soluble and therefore might be dangerous. This is an aspect which cannot be commented upon at the moment through lack of knowledge, it is however an important aspect that must be studied. The results may well explain the "anomalies" which so often confront the investigator. The effect of deoxidation practice on "weldability" should also be studied in view of the interrelated effects of manganese, sulphur, silicon, aluminium and oxygen.

Conclusions

(1) The most harmful elements causing cracking in ferritic steels are, in decreasing order of potency:
S, B, (Ce?), P, As, Nb, Sn, Zr, and Ta

(2) Carbon and copper in addition to the elements in (1) are harmful in austenitic steels.

(3) Carbon and nickel are crack promoters as they are austenisers while molybdenum, vanadium and chromium, being ferrite formers, are beneficial.

(4) Sulphide formers, lanthanum, cerium and manganese are beneficial. Lanthanum should be the most useful addition.

(5) Deoxidisers such as aluminium and silicon can be harmful when added to a high oxygen, low manganese mild steel.

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B.W.R.A. REPORT

Influence of Artificially Induced Residual Stresses on Fatigue Strength of Load-Carrying Fillet Welded Joints in Mild Steel

By T. R. Gurney, M.A., A.M.I.C.E.

The report contains details of fatigue tests carried out to determine whether spot heating and local compression, which are known to produce a considerable increase in the fatigue strength of specimens containing non-load-carrying welds, also improve the fatigue behaviour of specimens with load-carrying welds. It is shown that strength can be increased by about 100% at 2×10^6 cycles.

PREVIOUS work by Puchner,[1] whose results were confirmed by the present author, and Nekanda Trepka[2] showed that a large increase in the fatigue strength of specimens containing discontinuous longitudinal welds could be obtained by spot heating to introduce compressive residual stresses at the notch at the end of the weld. Further, it has been shown[3] that a similar increase in strength is obtainable by introducing the residual stresses mechanically by local compression. All these results, however, were obtained on specimens in which the welds were basically non-load-carrying.

In view of this it was decided to carry out further tests to determine whether similar increases could be produced in the fatigue strength of specimens containing load-carrying fillet welds, *i.e.*, specimens in which the load was transferred through the weld in shear. The results of this programme, which included tests under both pulsating tension and alternating load cycles, are given in this report. In the course of the investigation it became necessary to examine, in greater detail than had been intended, the effect of certain aspects of specimen geometry on fatigue strength, and the results of this subsidiary investigation are also given.

Test Specimens

Material

The material used in this programme of tests was $\frac{1}{2}$ in. and $\frac{3}{8}$ in. thick mild steel plate which complied essentially with BS.15, Structural Steel for Bridges and General Building Construction. The $\frac{1}{2}$ in. thick material was a basic open-hearth semi-killed steel; the origin of the $\frac{3}{8}$ in. thick plate is not known. The chemical analyses and mechanical properties of the materials are shown in Tables I and II, from which it will be seen that there was little difference between the properties of the two steels.

Specimen form and manufacture

Two basic types of specimen, each containing a fillet welded joint through which the total load in the specimen was carried in shear on the weld throats, were used in the test programme. One of these, which is hereafter referred to as the 'cover plate type' of specimen (Fig. 1), was designed to investigate the effect of spot heating. The other was designed for treatment by local compression and will be referred to as the 'egg box type' of specimen (Fig. 3).

The basic characteristics of any particular series of specimens are designated by identification letters, as follows:

The first letter denotes the type of specimen: thus *C*—cover plate type, *E*—egg box type.

The second letter, *N*, shows that the specimen was not stress relieved.

The figure 2 applied to cover plate type specimens indicates $\frac{1}{2}$ in. thick cover plates, all specimens without this figure having $\frac{3}{8}$ in. thick cover plates.

The letter *A* signifies groups tested under an alternating cycle, while all other groups were tested in pulsating tension with a lower limit stress of zero.

The post weld treatments are defined by the following letters:

P—locally pressed
O—main and cover plates spot heated in resistance welding machine
OX—main and cover plates spot heated by oxyacetylene torches
M—main plates only spot heated (by oxyacetylene torches)
Q—cover plates only spot heated (by oxyacetylene torches)

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Mr. Gurney is a Principal Research Officer with the Association.

Table I

Chemical analysis of BS. 15 plate material

Element	*½ in. Plate*	*¾ in. Plate*
Carbon	0·15	0·19
Manganese	0·74	0·66
Silicon	0·035	0·02
Sulphur	0·038	0·05
Phosphorus	0·028	0·03

Table II

Mechanical properties of BS. 15 plate material

	Yield Stress, tons/sq.in.	*U.T.S., tons/sq.in.*	*Elong. on 4 √A, %*	*Red. in Area, %*
½ in. plate	17·8	28·8	35	65·4
¾ in. plate	18·0	29·0	32·8	

Table III

Spot heating conditions

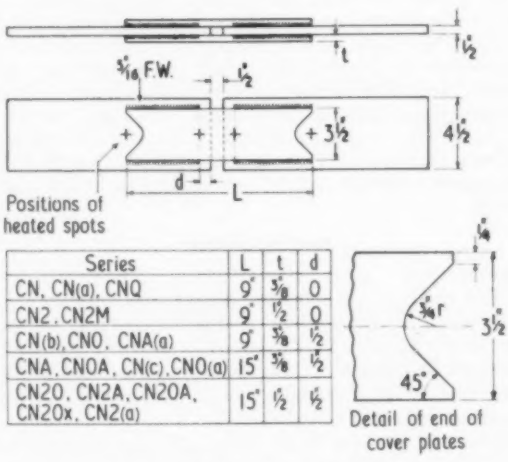
Specimen	*Electrode dia., in.*	*Load, lb.*	*Current, kA*	*Heating Time, sec*	*Transformer Tap*
½ in. main plate	⅜	500	8·2	55	M3. 230
¾ in. cover plates	½	500	10	50	M1. 230
½ in. cover plates	⅜	500	11	50	H3. 230

It should be noted that both basic types of specimen were symmetrical about their longitudinal centre lines so as to eliminate bending effects as far as possible, and also that both contained only longitudinal welds. It has been shown previously that, for the methods of treatment under consideration to be effective, it is essential for the 'notches', at which fatigue failure may be initiated, to be localised, and this can only be achieved by using longitudinal welds, which can be expected to produce failure from the weld ends. In the cover plate type specimens the ends of the cover plates were cut away so as to enable the local heating to be applied directly to the main plate on the line connecting the weld ends.

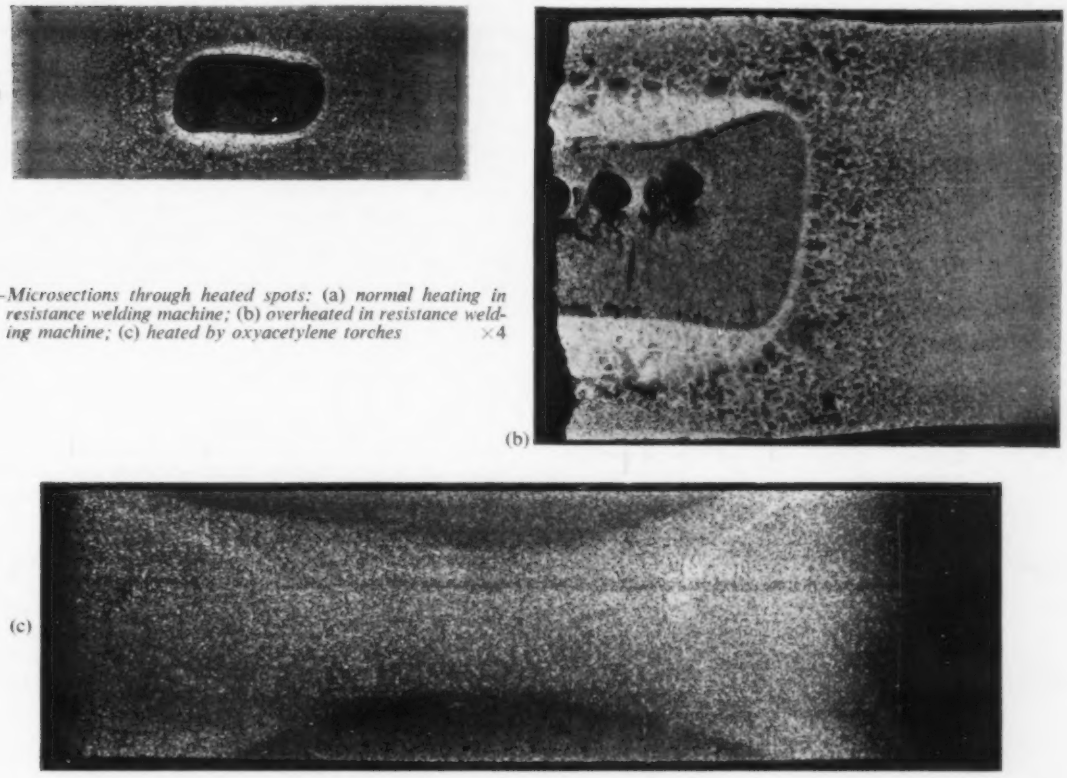
All the plates used were machine gas cut and the specimens were fabricated so that the direction of stressing was parallel to the rolling direction. Welding was carried out in the downhand position using rutile electrode E 217. In most cases the fillet welds were made in two halves starting at the ends of the weld runs and joining in the middle, so as to avoid as far as possible the effects of end craters. However, for comparison, some test series were welded in one direction only. Thus three series of cover plate specimens, *CN(a)* *CN(b)* and *CNQ*, were welded from the centre line towards the ends, while three series of similar specimens, *CN*, *CN2*, and *CN2M* were welded in the opposite direction. After fabrication the corners of the cross-sections were finished by hand to a radius of approximately $\frac{1}{16}$ in. to minimise the possibility of failure being initiated at a gas cut edge. None of the specimens was stress relieved.

Spot heating of the cover plate specimens forming series, *CNO*, *CNO(a)*, *CN2O*, *CNOA* and *CN2OA* was applied in the positions indicated in Fig. 1 by means of a resistance welding machine with ⅜ in. dia electrodes, the pressure between the electrodes being 500 lb. There was some variation of transformer tap, and consequently of heating time, between the different groups and even between different specimens in the same group. The object throughout was to obtain a heated spot of constant size, with a diameter as defined by the 280° isotherm of approximately 2½ in. By trial and error the heating conditions given in Table III were found to be satisfactory. A typical macrosection of the main plate of a specimen heated by this process is shown in Fig. 2*a*. It will be seen that the material in the centre of the single main plates became molten during heating, and that the resulting nugget contained pores and hot cracks, but this did not occur when heating the cover plates. Examination of the macrosections also enables further deductions to be made about the temperature distribution around the heated spot. Thus, in the main plate, the edge of the nugget (radius 0·19 in.) can be taken as approximately 1500°C. and the position of the eutectoid point shows that a temperature of 723°C. was reached at 0·52 in. radius. It is interesting to note that at this radius the temperature field was uniformly distributed through the plate thickness. Similarly, examination of a ¾ in. thick cover plate showed that a temperature of 895°C. was reached at a radius of approx. 0·31 in. and 723°C. at 0·46 in. radius. In Fig. 2*b* a macrosection is shown of one of four specimens in series *CN2OA*, which were overheated and resulted in failure from the heated spot. These results will be considered later.

Specimens in series *CN2OX*, *CNQ* and *CN2M* were also spot heated, but this was carried out by means of two oxyacetylene torches fitted with ½ in. dia. high output nozzles using the same technique as has been described elsewhere.[2] As far as possible the heated spots were made the same size as those produced in the resistance welding machine. In series *CN2OX* both the cover plates and the main plates were spot heated, but in series *CNQ* only the cover plates and in series



1—*Details of cover plate type specimens*



2—*Microsections through heated spots:* (a) *normal heating in resistance welding machine;* (b) *overheated in resistance welding machine;* (c) *heated by oxyacetylene torches* ×4

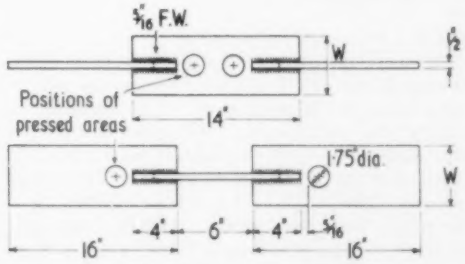
CN2M only the main plates. In the latter two series the object was to obtain failures in the main plate when the cover plates were $\frac{3}{8}$ in. thick (normally failure was to be expected in the cover plates) and in the cover plates when they were $\frac{1}{2}$ in. thick (in which case failure occurred normally in the main plate). As will be shown later these efforts proved successful. A macro-section of a typical specimen heated by oxyacetylene torches is shown in Fig. 2*c*.

Local compression treatment was applied to the egg box type specimens in series *ENP* and *ENPA* with $1\frac{3}{4}$ in. diameter dies in the positions indicated in Fig. 3, the compression being carried out in a static testing machine. The total load applied was 280,000 lb (equivalent to an average compressive stress under the dies of 51·5 tons/sq.in.) for 30 sec, and this resulted in an average residual indentation of 0·007 in. in each face (*i.e.*, a strain in the direction of plate thickness of 2·8%).

Method of Testing

All the specimens were tested under axial loading, the applied load cycle being either pulsating tension, with a lower limit of zero, or alternating ($f_{min} = -f_{max}$). The load cycle used for any particular specimen is shown in Tables IV—XVI. The majority of the tests were carried out in a 60 ton Schenck machine at 2,000 c/min, but some specimens were tested in a 100 ton Losenhausen hydraulic machine with a testing frequency in the range 200–400 c/min. The machine used to test any particular specimen is indicated in Tables IV–XV, but it has been found that there is no significant difference between the results obtained on the two machines.

The testing procedure was designed to establish S/N curves for the range of endurance values from approximately 100,000 to 2,000,000 cycles. However, in many of the test groups only a comparatively small number of specimens were tested, and few attempts were made to produce unbroken specimens after 2,000,000 cycles. Hence the absolute values of fatigue strength recorded in this work should be regarded as slightly optimistic, but it is considered that certain qualitative conclusions can be drawn from the work. Any specimen which survived the fatigue test was tested statically to failure to ensure that no fatigue cracks, at least of macroscopic dimensions, were present, and in cases where cracks were found the specimen has been regarded as 'failed'. In all other cases the criterion of failure was taken as complete rupture of the specimen.



3—*Details of egg box type specimens*

Table IV

Summary of results for series CN, CN(a), CN(b) and CN(c)
(Specimens with ½ in. thick cover plates, as-welded)

Specimen Ref. No.	Testing Machine	Stress Range, tons/sq.in. Main Plate	Stress Range, tons/sq.in. Cover Plates	Cycles to Failure $\times 10^6$	Remarks
CN/1	Schenck	0—12	0—10·3	0·067	Failed from weld ends in cover plates
2	Schenck	0— 9	0— 7·7	0·175	
3	Los	0— 7·5	0— 6·4	0·409	
4	Los	0— 6·5	0— 5·6	0·687	
5	Schenck	0— 4	0— 3·4	1·879	
6	Los	0— 3·5	0— 3	2·101	2 small cracks in cover plates found in static test
CN(a)/1	Schenck	0—10	0— 8·5	0·114	Failed from weld ends in cover plates
2	Schenck	0— 7	0— 6	0·314	
3	Schenck	0— 5·5	0— 4·7	1·185	
4	Schenck	0— 4·5	0— 3·8	2·920	
CN(b)/1	Schenck	0—10·5	0— 9	0·105	Failed from weld ends in cover plates
2	Schenck	0— 8	0— 6·8	0·269	
3	Schenck	0— 5	0— 4·3	0·953	
4	Schenck	0— 4·5	0— 3·8	2·879	
CN(c)/1	Schenck	0—11	0— 9·4	0·135	Failed from weld ends in cover plates
2	Schenck	0— 9	0— 7·7	0·272	
3	Los	0— 6·5	0— 5·6	1·074	
4	Schenck	0— 5·5	0— 4·7	1·654	
5	Schenck	0— 5	0— 4·3	2·107	

Results and Discussion of Tests

It is convenient to consider the results obtained in two parts, as follows:

(i) Tests on cover plate type specimens in the as-welded condition to investigate the effect of differences in specimen geometry. This was in fact a subsidiary investigation carried out in the process of determining control values for the second part
(ii) Tests to determine the effects of spot heating and local compression.

Effect of specimen geometry

In the design of the cover plate type specimens there were three main variables:

(*a*) The direction of welding
(*b*) The length of welding and its disposition relative to the attached plates
(*c*) The thickness of cover plate.

Influence of direction of welding

A direct comparison between specimens welded in different directions is given by series *CN* and *CN(a)*.

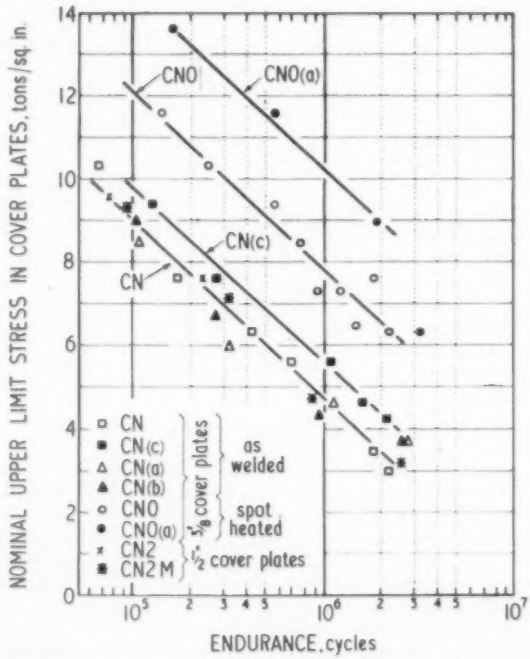
In series *CN* the welds were made inwards, towards the centre of the specimens, while in series *CN(a)* the welds were made in the outward direction. In all other respects the two groups were identical. The results are shown in Table IV and have been plotted in Fig. 4. from which it can be seen that there was no significant difference in fatigue strength between the two groups, all specimens failing through the cover plates from the weld ends. In view of these results it is not considered important that, in the course of the test programme, groups welded in different directions have been compared with each other. However, in view of the possibility that the presence of end craters might influence fatigue strength, the fillet welds of most of the test series were deposited in two half length runs starting from the two ends so as to leave the crater near the centre of the weld length.

To investigate the possible influence of the position of the weld end relative to the (central) end of the main plate, in series *CN(b)* a few specimens were tested which were welded from the centre outwards and had the start of the weld removed from this edge of the plate, as shown in Fig. 1. The results obtained are shown in Table IV and Fig. 4 and are directly comparable with those obtained for series *CN(a)*. It will be seen that there is no significant difference between the results for the two groups, but the majority of test series were welded in a similar manner to *CN(b)* since this was a slightly easier method and was expected therefore to result in more consistent welds.

Influence of weld length

Weld length, which was found to be a very significant variable, was investigated in some detail, although the number of specimens subjected to fatigue tests was necessarily limited. In spite of this, the consistent lack of scatter of the test results enables reasonable conclusions to be drawn. The results for the following test series can be compared:

CN and *CN(c)*	(Table IV, Fig. 4)
CN2 and *CN2(a)*	(Table VI, Fig. 5)
CNO and *CNO(a)*	(Table V, Fig. 4)
CNA and *CNA(a)*	(Table VIII, Fig. 6)



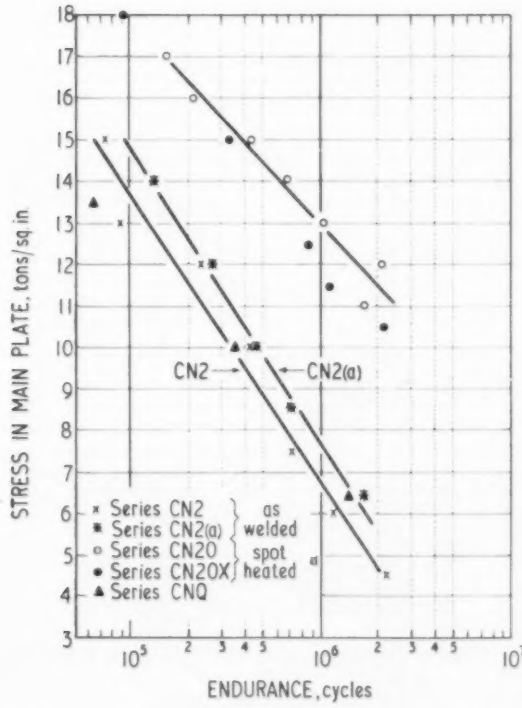
4—*Fatigue test results for cover plate type specimens failing in the cover plates. (Pulsating tension cycle)*

Table V

Summary of results for series CNO, CNO(a) and CNQ

(Specimens with $\frac{3}{4}$ in. thick cover plates, spot heated)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CNO/1	Los	0—13·5	0—11·6	0·147	Main failure from weld ends through cover plates. Small cracks at weld ends in main plate. One crack in weld throat at 'central' end and 2 at 'gripped' ends.
2	Schenck	0—12	0—10·3	0·245	Failure from end of weld in cover plate and 4 cracks in weld throats (1 'central' end and 3 'gripped' end)
3	Schenck	0—11	0— 9·4	0·570	Failure from weld end through cover plate
4	Schenck	0—10	0— 8·5	0·734	Failure from end of weld in cover plate.
5	Schenck	0— 9	0— 7·7	1·873	Failure from end of weld through cover plate and crack in weld throat ('gripped' end)
6	Schenck	0— 8·5	0— 7·3	0·918	Failure from weld end through cover plate and 2 cracks in weld throat ('central' ends)
7	Schenck	0— 8·5	0— 7·3	1·293	Failure from weld end in cover plate and crack in weld throat ('gripped' end)
8	Schenck	0— 7·7	0— 6·5	1·542	Failure from weld end in cover plate
9	Los	0— 7·5	0— 6·4	2·227	Small cracks in cover plates at weld ends found in static test. Also transverse crack through weld at crater and crack in weld throat at weld end.
CNO(a)/1	Los	0—16	0—13·7	0·164	Failure from end of weld through cover plate
2	Schenck	0—13·5	0—11·6	0·568	
3	Schenck	0—10·5	0— 9	1·927	
4	Los	0— 7·5	0— 6·4	3·101	Very small cracks at weld ends in cover plate found in static test
CNQ/1	Los	0—13·5	0—11·6	0·066	Failure at weld end in main plate
2	Los	0—10	0— 8·5	0·356	Failure at weld end in main plate and cracks in 3 weld throats
3	Los	0— 6·5	0— 5·6	1·470	Failure at weld end in main plate



5—Fatigue test results for cover plate type specimens failing in the main plates. (Pulsating tension cycle)

The only difference between the two groups forming each pair was the length of the welds, and a comparison of the fatigue strengths obtained at 10^5 and 2×10^6 cycles for the first two pairs is:

Series	*Weld Length, in.*	*Point of Failure*	*Fatigue Strength, tons/sq.in* 10^5	2×10^6	*Theoretical S.C.F. Main Plate*	*Cover Plate*
Stress in cover plate						
CN	4·25	Through cover plates	9·0	3·5	(1·81)	2·07
CN(c)	6·75		9·8	4·4	(1·46)	1·58
Stress in main plate						
CN2	4·25	Through main plate	13·8	4·6	1·91	(1·95)
CN2(a)	6·75		14·9	5·6	1·52	(1·51)

In this table are also shown the theoretical stress concentration factors computed by the method developed by Palmer;[4] it will be seen that these provide at least a qualitative explanation of the greater fatigue strengths obtained with longer welds. A similar trend can be observed when the other two pairs of groups are compared, but both series *CNO(a)* and *CNA(a)* consisted of only 4 specimens so that it is not possible to quote accurate fatigue strengths. In Fig. 6, where groups *CNA* and *CNA(a)* are compared, the open triangles indicate specimens in series *CNA* which failed through the main plate instead of through the cover plate. These points therefore represent a lower limit to the fatigue strength of the cover plates. All the specimens in series *CNA(a)* failed from the weld ends through the cover plates. It can be seen that the approximate fatigue strength at 2×10^6 cycles for

Table VI

Summary of results for series CN2 and CN2(a)

(Specimens with ½ in. thick cover plates, as-welded)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CN2/1	Los	0—15	0—9·65	0·074	Failure on weld throat. Two small cracks in main plate and three in cover plates
2	Los	0—13	0—8·4	0·089	Failure from end of weld through main plate
3	Los	0—12	0—7·7	0·230	Failure at weld ends in cover plates and main plate and also in a weld throat
4	Los	0—10	0—6·4	0·419	Failure at weld ends through main plate
5	Los	0— 7·5	0—4·8	0·727	
6	Schenck	0— 6	0—3·8	1·146	
7	Los	0— 4·5	0—2·9	2·226	Four cracks at weld ends in main plate found in static test
CN2(a)/1	Los	0—14	0—9·0	0·140	Failed from end of weld through main plate
2	Los	0—12	0—7·7	0·272	
3	Los	0—10	0—6·4	0·428	
4	Los	0— 8·5	0—5·5	0·702	
5	Los	0— 6·5	0—4·2	1·680	

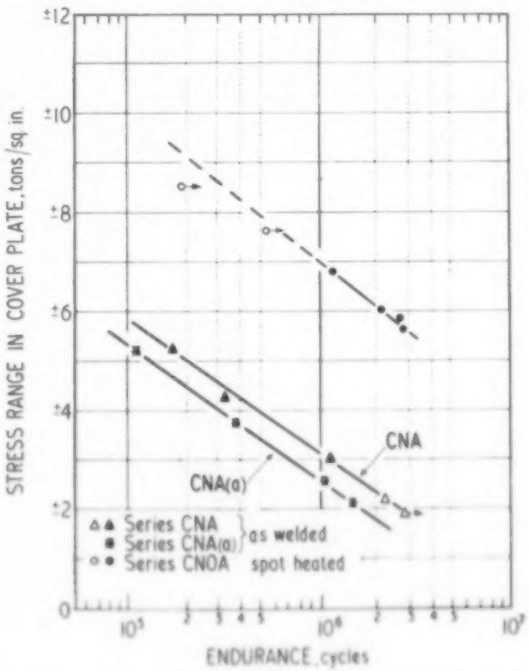
specimens with short welds (series *CNA(a)*) was ±1·8 tons/sq.in. and that this was increased to approximately ±2·3 tons/sq.in. by increasing the weld length, these being the nominal average stresses in the cover plates. In the case of groups *CNO* and *CNO(a)* all the main failures were initiated at the weld ends and propagated through the cover plates. It can be seen from Fig. 4 that at 2×10^6 cycles, series *CNO*, with short welds, gave a fatigue strength of 6·6 tons/sq.in. in the cover plates, but that this was increased to approx. 9 tons/sq.in. for series *CNO(a)*. In drawing the S/N curve for this group the result for specimen *CNO(a)*/4 has been ignored, even though the specimen technically had failed. The fatigue cracks discovered by the static test were extremely small and there is little doubt that, if the specimen had been allowed to run to final rupture, it would have survived for a considerable time, because the rate of crack propagation in spot heated specimens is known to be slow.

To sum up, the effect of increasing weld length was to produce an increase of fatigue strength of approximately 25% for as-welded specimens and 35% for spot heated specimens.

Influence of cover plate thickness

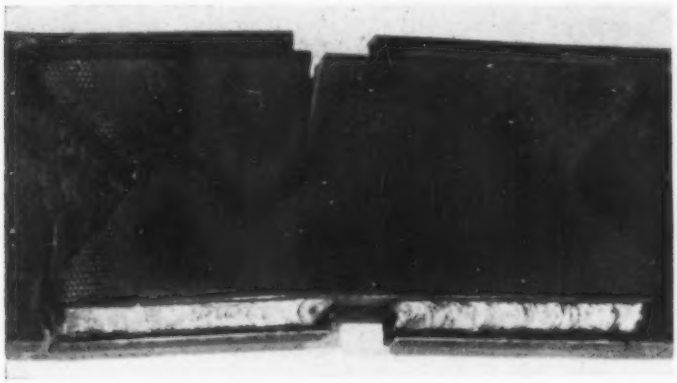
The effect of cover plate thickness on the fatigue strength of as-welded specimens tested in pulsating tension can be seen by comparing the results for series *CN* and *CN2* which were otherwise identical.(Tables IV and VI, Figs. 4 and 5). In series *CN* all the specimens failed through the cover plates, the cracks in specimen *CN*/6 being discovered in the subsequent static test. In series *CN2*, however, all the specimens developed cracks at the ends of the welds in the main plate, although specimens *CN2*/1 and 3 also had cracks in the cover plates. These two results have also been plotted in Fig. 4 and it will be seen that they lie close to the S/N curve for series *CN*. A photograph of the failure in specimen *CN2*/3 is shown in Fig. 7. In specimen *CN2*/7 the fatigue cracks were found in the subsequent static test after the specimen had survived $2{\cdot}2 \times 10^6$ cycles at 0—4·5 tons/sq.in. The fatigue strength at 2×10^6 cycles is shown by these results to be approximately 3·5 tons/sq.in. in the cover plates and 4·6 tons/sq.in. in the main plate. It should be noted that the former figure was confirmed by tests on series *CN2M* which was identical with series *CN2* but was spot heated to prevent failure of the main plates. The results for this series are given in Table VII and have been plotted in Fig. 4.

A similar comparison can be made between series *CNA* and *CN2A*, which were both tested on an alternating cycle. The results obtained are given in Tables VIII and X and are shown graphically in Figs. 6 and 8.



6—*Fatigue test results for cover plate type specimens failing in the cover plates. (Alternating cycle)*

All the specimens in series *CN2A* failed from the weld ends through the main plate, the failure in specimen *CN2A*/6, tested at ± 2 tons/sq.in., being found in the static test after the specimen had survived $4{\cdot}6 \times 10^6$ load cycles under full load. However, in series *CNA* the position of failure was randomly distributed between the main plate and cover plates and the results therefore have been plotted on both Fig. 6 and



7—*Failure of specimen* CN2/3

Table VII

Summary of results for series CN20, CN20X and CN2M

(Specimens with $\frac{1}{2}$ in. thick cover plates, spot heated)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CN2O/1	Schenck	0—17	0—10·9	0·161	Failed in main plate from weld root approx. $\frac{3}{4}$ in. inside end of weld
2	Los	0—16	0—10·3	0·207	
3	Schenck	0—15	0— 9·6	0·436	
4	Los	0—14	0— 9·0	0·681	
5	Schenck	0—13	0— 8·4	1·031	Failed from end of weld in main plate
6	Schenck	0—12	0— 7·7	2·101	Failed in main plate from weld root approx. $\frac{3}{4}$ in. inside end of weld
7	Schenck	0—11	0— 7·1	1·769	
CN2M/1	Schenck	0—14·5	0— 9·3	0·097	Failed from end of weld in cover plates
2	Schenck	0—11	0— 7·1	0·312	
3	Los	0— 7·5	0— 4·8	0·873	
4	Schenck	0— 5·0	0— 3·2	2·558	
CN2OX/1	Los	0—18	0—11·6	0·091	Failed from ends of welds in main plates
2	Schenck	0—15	0— 9·6	0·334	
3	Schenck	0—12·5	0— 8·0	0·858	
4	Schenck	0—11·5	0— 7·4	1·131	Failed from end of weld in main plate and from surface in heated spot
5	Los	0—10·5	0— 6·7	2·155	Small cracks at weld ends in main plates found in static test

Table VIII

Summary of results for series CNA and CNA(a)

(Specimens with $\frac{1}{2}$ in. thick cover plates, as-welded)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CNA/1	Schenck	±6·0	±5·15	0·170	Failed from end of weld through cover plated
2	Los	±5·0	±4·3	0·312	
3	Schenck	±3·5	±3·0	1·192	Failed from end of weld through main plate. Two small cracks at weld ends in cover plates found ultrasonically
4	Schenck	±3·0	±2·6	1·066	Failed at end of weld through main plate
5	Schenck	±2·5	±2·15	2·142	As for CNA/3
6	Schenck	±2·25	±1·9	2·825	Failed at end of weld through main plate
CNA(a)/1	Schenck	±6·0	±5·15	0·115	Failed from ends of welds through cover plate
2	Los	±4·5	±3·85	0·361	
3	Schenck	±3·0	±2·6	1·054	
4	Schenck	±2·5	±2·1	1·513	

C = 0·12 max.
Mn = 0·50 min.
S = 0·035 max.
P = 0·040 max.
Mn/S ratio 14·3 min.

This should be suitable for welds made under low restraint. For more severe conditions, and where good impact properties are desired the manganese should be increased, 0·8–1% being sufficient in the majority of cases.

Preventing cracking in low alloy steels

Ideally, the most suitable weld metal composition would be one initially solidifying mainly as ferrite. The ratio of austenisers and ferritisers should thus be balanced to obtain this. Where this is impossible, then the sulphur and phosphorus should be reduced to a very low level; in some applications where high strength and good notch toughness is desired the amounts should not exceed 0·005% each, the Mn/S ratio being greater than about 70. Some additional benefit, particularly in regard to weld metal properties, may be obtained possibly by also reducing the arsenic and antimony content, though experimental evidence for this has not yet been obtained as far as the author is aware.

It is not possible to recommend weld metal compositions which should be free from cracking as suitable data do not yet exist. However it is obvious that high carbon and nickel* contents are harmful unless sulphur and phosphorus contents are strictly controlled. Molybdenum, vanadium and chromium should be beneficial, as should also be manganese, lanthanum, and cerium in controlled amounts. Boron should be harmful.

* It would be interesting to know the solubility of carbon in an alloy containing 3–4%Ni. The effect of carbon at this level may not be additive, in fact the solubility of carbon might be relatively high.

Finally, it is generally recognised that the effect of minor elements becomes more important as the strength of the material is increased; the steels discussed here should not be an exception to this rule. Thus in ultra-high strength steels it is likely that the permissible limits of sulphur and phosphorus will decrease; also it is likely that elements such as arsenic and tin will assume more importance.

Normally class 'O' elements should not be harmful; it is a possibility however that in certain alloys containing perhaps a high proportion of a major element, these insoluble elements will become slightly soluble and therefore might be dangerous. This is an aspect which cannot be commented upon at the moment through lack of knowledge, it is however an important aspect that must be studied. The results may well explain the "anomalies" which so often confront the investigator. The effect of deoxidation practice on "weldability" should also be studied in view of the interrelated effects of manganese, sulphur, silicon, aluminium and oxygen.

Conclusions

(1) The most harmful elements causing cracking in ferritic steels are, in decreasing order of potency:
S, B, (Ce?), P, As, Nb, Sn, Zr, and Ta

(2) Carbon and copper in addition to the elements in (1) are harmful in austenitic steels.

(3) Carbon and nickel are crack promoters as they are austenisers while molybdenum, vanadium and chromium, being ferrite formers, are beneficial.

(4) Sulphide formers, lanthanum, cerium and manganese are beneficial. Lanthanum should be the most useful addition.

(5) Deoxidisers such as aluminium and silicon can be harmful when added to a high oxygen, low manganese mild steel.

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B.W.R.A. REPORT

Influence of Artificially Induced Residual Stresses on Fatigue Strength of Load-Carrying Fillet Welded Joints in Mild Steel

By T. R. Gurney, M.A., A.M.I.C.E.

The report contains details of fatigue tests carried out to determine whether spot heating and local compression, which are known to produce a considerable increase in the fatigue strength of specimens containing non-load-carrying welds, also improve the fatigue behaviour of specimens with load-carrying welds. It is shown that strength can be increased by about 100% at 2×10^6 cycles.

PREVIOUS work by Puchner,[1] whose results were confirmed by the present author, and Nekanda Trepka[2] showed that a large increase in the fatigue strength of specimens containing discontinuous longitudinal welds could be obtained by spot heating to introduce compressive residual stresses at the notch at the end of the weld. Further, it has been shown[3] that a similar increase in strength is obtainable by introducing the residual stresses mechanically by local compression. All these results, however, were obtained on specimens in which the welds were basically non-load-carrying.

In view of this it was decided to carry out further tests to determine whether similar increases could be produced in the fatigue strength of specimens containing load-carrying fillet welds, *i.e.*, specimens in which the load was transferred through the weld in shear. The results of this programme, which included tests under both pulsating tension and alternating load cycles, are given in this report. In the course of the investigation it became necessary to examine, in greater detail than had been intended, the effect of certain aspects of specimen geometry on fatigue strength, and the results of this subsidiary investigation are also given.

Test Specimens

Material

The material used in this programme of tests was $\frac{1}{2}$ in. and $\frac{3}{8}$ in. thick mild steel plate which complied essentially with BS.15, Structural Steel for Bridges and General Building Construction. The $\frac{1}{2}$ in. thick material was a basic open-hearth semi-killed steel; the origin of the $\frac{3}{8}$ in. thick plate is not known. The chemical analyses and mechanical properties of the materials are shown in Tables I and II, from which it will be seen that there was little difference between the properties of the two steels.

Specimen form and manufacture

Two basic types of specimen, each containing a fillet welded joint through which the total load in the specimen was carried in shear on the weld throats, were used in the test programme. One of these, which is hereafter referred to as the 'cover plate type' of specimen (Fig. 1), was designed to investigate the effect of spot heating. The other was designed for treatment by local compression and will be referred to as the 'egg box type' of specimen (Fig. 3).

The basic characteristics of any particular series of specimens are designated by identification letters, as follows:

The first letter denotes the type of specimen: thus *C*—cover plate type, *E*—egg box type.

The second letter, *N*, shows that the specimen was not stress relieved.

The figure 2 applied to cover plate type specimens indicates $\frac{1}{2}$ in. thick cover plates, all specimens without this figure having $\frac{3}{8}$ in. thick cover plates.

The letter *A* signifies groups tested under an alternating cycle, while all other groups were tested in pulsating tension with a lower limit stress of zero.

The post weld treatments are defined by the following letters:

P—locally pressed
O—main and cover plates spot heated in resistance welding machine
OX—main and cover plates spot heated by oxyacetylene torches
M—main plates only spot heated (by oxyacetylene torches)
Q—cover plates only spot heated (by oxyacetylene torches)

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Mr. Gurney is a Principal Research Officer with the Association.

Table I

Chemical analysis of BS. 15 plate material

Element	*½ in. Plate*	*⅜ in. Plate*
Carbon	0·15	0·19
Manganese	0·74	0·66
Silicon	0·035	0·02
Sulphur	0·038	0·05
Phosphorus	0·028	0·03

Table II

Mechanical properties of BS. 15 plate material

	Yield Stress, tons/sq.in.	*U.T.S., tons/sq.in.*	*Elong. on 4 √A, %*	*Red. in Area, %*
½ in. plate	17·8	28·8	35	65·4
⅜ in. plate	18·0	29·0	32·8	

Table III

Spot heating conditions

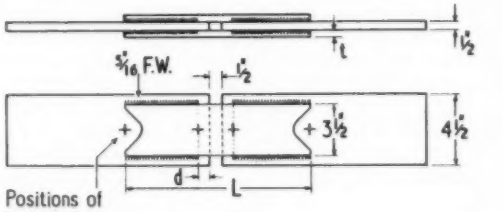
Specimen	*Electrode dia., in.*	*Load, lb.*	*Current, kA*	*Heating Time, sec*	*Transformer Tap*
½ in. main plate	⅝	500	8·2	55	M3. 230
⅜ in. cover plates	⅝	500	10	50	M1. 230
½ in. cover plates	⅝	500	11	50	H3. 230

It should be noted that both basic types of specimen were symmetrical about their longitudinal centre lines so as to eliminate bending effects as far as possible, and also that both contained only longitudinal welds. It has been shown previously that, for the methods of treatment under consideration to be effective, it is essential for the 'notches', at which fatigue failure may be initiated, to be localised, and this can only be achieved by using longitudinal welds, which can be expected to produce failure from the weld ends. In the cover plate type specimens the ends of the cover plates were cut away so as to enable the local heating to be applied directly to the main plate on the line connecting the weld ends.

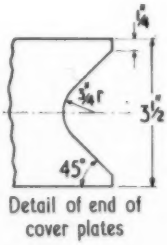
All the plates used were machine gas cut and the specimens were fabricated so that the direction of stressing was parallel to the rolling direction. Welding was carried out in the downhand position using rutile electrode E 217. In most cases the fillet welds were made in two halves starting at the ends of the weld runs and joining in the middle, so as to avoid as far as possible the effects of end craters. However, for comparison, some test series were welded in one direction only. Thus three series of cover plate specimens, *CN(a)* *CN(b)* and *CNQ*, were welded from the centre line towards the ends, while three series of similar specimens, *CN*, *CN2*, and *CN2M* were welded in the opposite direction. After fabrication the corners of the cross-sections were finished by hand to a radius of approximately $\frac{1}{16}$ in. to minimise the possibility of failure being initiated at a gas cut edge. None of the specimens was stress relieved.

Spot heating of the cover plate specimens forming series, *CNO*, *CNO(a)*, *CN2O*, *CNOA* and *CN2OA* was applied in the positions indicated in Fig. 1 by means of a resistance welding machine with ⅝ in. dia electrodes, the pressure between the electrodes being 500 lb. There was some variation of transformer tap, and consequently of heating time, between the different groups and even between different specimens in the same group. The object throughout was to obtain a heated spot of constant size, with a diameter as defined by the 280° isotherm of approximately 2½ in. By trial and error the heating conditions given in Table III were found to be satisfactory. A typical macrosection of the main plate of a specimen heated by this process is shown in Fig. 2*a*. It will be seen that the material in the centre of the single main plates became molten during heating, and that the resulting nugget contained pores and hot cracks, but this did not occur when heating the cover plates. Examination of the macrosections also enables further deductions to be made about the temperature distribution around the heated spot. Thus, in the main plate, the edge of the nugget (radius 0·19 in.) can be taken as approximately 1500°C. and the position of the eutectoid point shows that a temperature of 723°C. was reached at 0·52 in. radius. It is interesting to note that at this radius the temperature field was uniformly distributed through the plate thickness. Similarly, examination of a ⅜ in. thick cover plate showed that a temperature of 895°C. was reached at a radius of approx. 0·31 in. and 723°C. at 0·46 in. radius. In Fig. 2*b* a macrosection is shown of one of four specimens in series *CN2OA*, which were overheated and resulted in failure from the heated spot. These results will be considered later.

Specimens in series *CN2OX*, *CNQ* and *CN2M* were also spot heated, but this was carried out by means of two oxyacetylene torches fitted with ½ in. dia. high output nozzles using the same technique as has been described elsewhere.[2] As far as possible the heated spots were made the same size as those produced in the resistance welding machine. In series *CN2OX* both the cover plates and the main plates were spot heated, but in series *CNQ* only the cover plates and in series

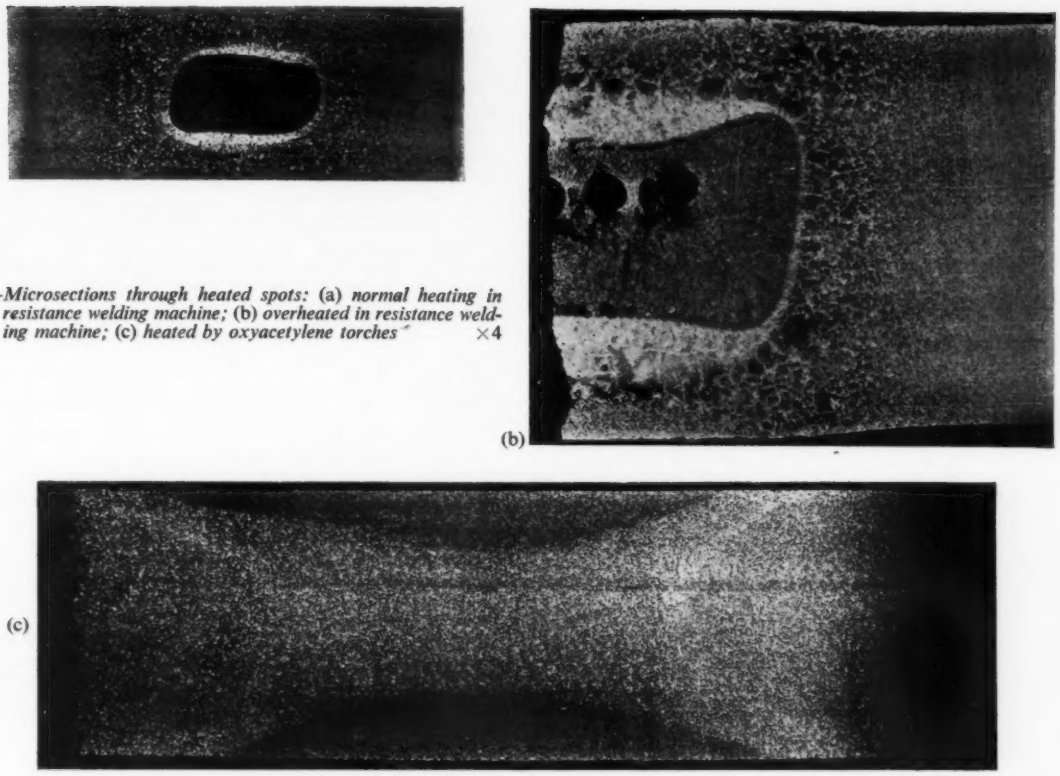


Series	L	t	d
CN, CN(a), CNQ	9"	3/8	0
CN2, CN2M	9"	½	0
CN(b), CNO, CNA(a)	9"	3/8	½
CNA, CNOA, CN(c), CNO(a)	15"	3/8	½
CN2O, CN2A, CN2OA, CN2Ox, CN2(a)	15"	½	½



Detail of end of cover plates

1—*Details of cover plate type specimens*



2—*Microsections through heated spots:* (a) *normal heating in resistance welding machine;* (b) *overheated in resistance welding machine;* (c) *heated by oxyacetylene torches* ×4

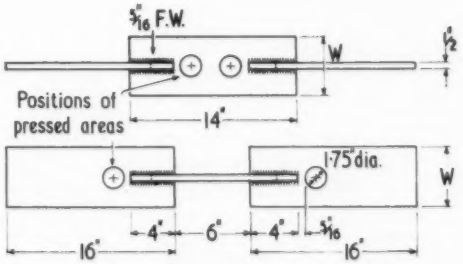
CN2M only the main plates. In the latter two series the object was to obtain failures in the main plate when the cover plates were $\frac{3}{8}$ in. thick (normally failure was to be expected in the cover plates) and in the cover plates when they were $\frac{1}{2}$ in. thick (in which case failure occurred normally in the main plate). As will be shown later these efforts proved successful. A macro-section of a typical specimen heated by oxyacetylene torches is shown in Fig. 2*c*.

Local compression treatment was applied to the egg box type specimens in series *ENP* and *ENPA* with $1\frac{3}{4}$ in. diameter dies in the positions indicated in Fig. 3, the compression being carried out in a static testing machine. The total load applied was 280,000 lb (equivalent to an average compressive stress under the dies of 51·5 tons/sq.in.) for 30 sec, and this resulted in an average residual indentation of 0·007 in. in each face (*i.e.*, a strain in the direction of plate thickness of 2·8%).

Method of Testing

All the specimens were tested under axial loading, the applied load cycle being either pulsating tension, with a lower limit of zero, or alternating ($f_{min} = -f_{max}$). The load cycle used for any particular specimen is shown in Tables IV—XVI. The majority of the tests were carried out in a 60 ton Schenck machine at 2,000 c/min, but some specimens were tested in a 100 ton Losenhausen hydraulic machine with a testing frequency in the range 200–400 c/min. The machine used to test any particular specimen is indicated in Tables IV–XV, but it has been found that there is no significant difference between the results obtained on the two machines.

The testing procedure was designed to establish S/N curves for the range of endurance values from approximately 100,000 to 2,000,000 cycles. However, in many of the test groups only a comparatively small number of specimens were tested, and few attempts were made to produce unbroken specimens after 2,000,000 cycles. Hence the absolute values of fatigue strength recorded in this work should be regarded as slightly optimistic, but it is considered that certain qualitative conclusions can be drawn from the work. Any specimen which survived the fatigue test was tested statically to failure to ensure that no fatigue cracks, at least of macroscopic dimensions, were present, and in cases where cracks were found the specimen has been regarded as 'failed'. In all other cases the criterion of failure was taken as complete rupture of the specimen.



3—*Details of egg box type specimens*

Table IV

Summary of results for series CN, CN(a), CN(b) and CN(c)
(Specimens with ⅜ in. thick cover plates, as-welded)

Specimen Ref. No.	Testing Machine	Stress Range, tons/sq.in. Main Plate	Stress Range, tons/sq.in. Cover Plates	Cycles to Failure $\times 10^6$	Remarks
CN/1	Schenck	0—12	0—10·3	0·067	Failed from weld ends in cover plates
2	Schenck	0— 9	0— 7·7	0·175	
3	Los	0— 7·5	0— 6·4	0·409	
4	Los	0— 6·5	0— 5·6	0·687	
5	Schenck	0— 4	0— 3·4	1·879	
6	Los	0— 3·5	0— 3	2·101	2 small cracks in cover plates found in static test
CN(a)/1	Schenck	0—10	0— 8·5	0·114	Failed from weld ends in cover plates
2	Schenck	0— 7	0— 6	0·314	
3	Schenck	0— 5·5	0— 4·7	1·185	
4	Schenck	0— 4·5	0— 3·8	2·920	
CN(b)/1	Schenck	0—10·5	0— 9	0·105	Failed from weld ends in cover plates
2	Schenck	0— 8	0— 6·8	0·269	
3	Schenck	0— 5	0— 4·3	0·953	
4	Schenck	0— 4·5	0— 3·8	2·879	
CN(c)/1	Schenck	0—11	0— 9·4	0·135	Failed from weld ends in cover plates
2	Schenck	0— 9	0— 7·7	0·272	
3	Los	0— 6·5	0— 5·6	1·074	
4	Schenck	0— 5·5	0— 4·7	1·654	
5	Schenck	0— 5	0— 4·3	2·107	

Results and Discussion of Tests

It is convenient to consider the results obtained in two parts, as follows:

(i) Tests on cover plate type specimens in the as-welded condition to investigate the effect of differences in specimen geometry. This was in fact a subsidiary investigation carried out in the process of determining control values for the second part
(ii) Tests to determine the effects of spot heating and local compression.

Effect of specimen geometry

In the design of the cover plate type specimens there were three main variables:

(*a*) The direction of welding
(*b*) The length of welding and its disposition relative to the attached plates
(*c*) The thickness of cover plate.

Influence of direction of welding

A direct comparison between specimens welded in different directions is given by series *CN* and *CN(a)*.

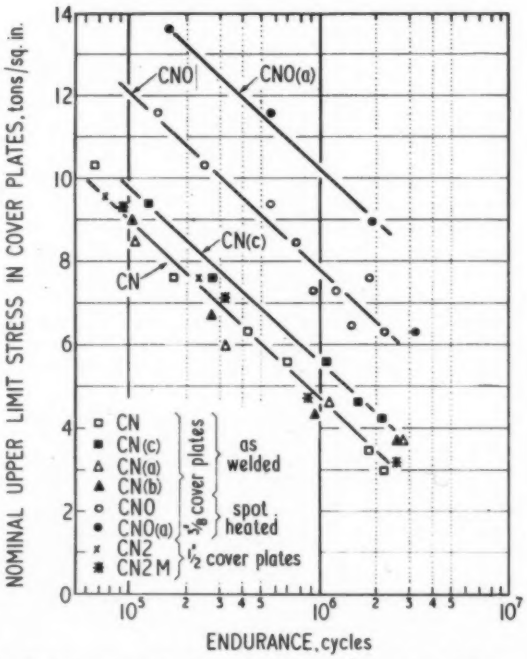
In series *CN* the welds were made inwards, towards the centre of the specimens, while in series *CN(a)* the welds were made in the outward direction. In all other respects the two groups were identical. The results are shown in Table IV and have been plotted in Fig. 4. from which it can be seen that there was no significant difference in fatigue strength between the two groups, all specimens failing through the cover plates from the weld ends. In view of these results it is not considered important that, in the course of the test programme, groups welded in different directions have been compared with each other. However, in view of the possibility that the presence of end craters might influence fatigue strength, the fillet welds of most of the test series were deposited in two half length runs starting from the two ends so as to leave the crater near the centre of the weld length.

To investigate the possible influence of the position of the weld end relative to the (central) end of the main plate, in series *CN(b)* a few specimens were tested which were welded from the centre outwards and had the start of the weld removed from this edge of the plate, as shown in Fig. 1. The results obtained are shown in Table IV and Fig. 4 and are directly comparable with those obtained for series *CN(a)*. It will be seen that there is no significant difference between the results for the two groups, but the majority of test series were welded in a similar manner to *CN(b)* since this was a slightly easier method and was expected therefore to result in more consistent welds.

Influence of weld length

Weld length, which was found to be a very significant variable, was investigated in some detail, although the number of specimens subjected to fatigue tests was necessarily limited. In spite of this, the consistent lack of scatter of the test results enables reasonable conclusions to be drawn. The results for the following test series can be compared:

CN and *CN(c)*	(Table IV, Fig. 4)
CN2 and *CN2(a)*	(Table VI, Fig. 5)
CNO and *CNO(a)*	(Table V, Fig. 4)
CNA and *CNA(a)*	(Table VIII, Fig. 6)



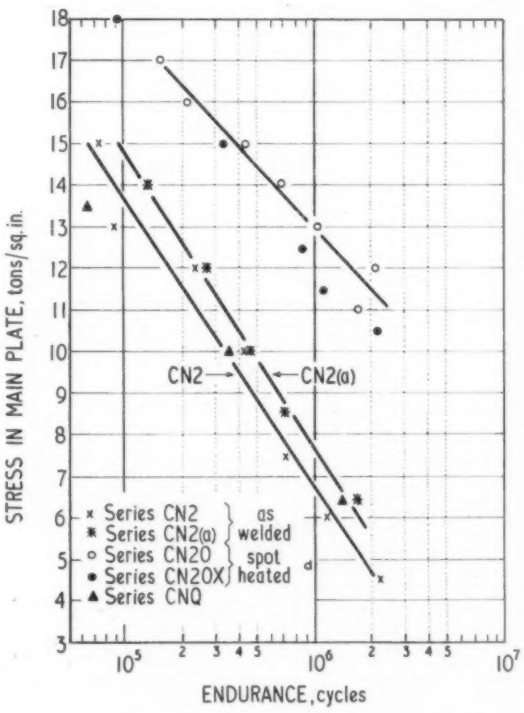
4—*Fatigue test results for cover plate type specimens failing in the cover plates. (Pulsating tension cycle)*

Table V

Summary of results for series CNO, CNO(a) and CNQ

(Specimens with $\frac{3}{8}$ in. thick cover plates, spot heated)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CNO/1	Los	0—13·5	0—11·6	0·147	Main failure from weld ends through cover plates. Small cracks at weld ends in main plate. One crack in weld throat at 'central' end and 2 at 'gripped' ends.
2	Schenck	0—12	0—10·3	0·245	Failure from end of weld in cover plate and 4 cracks in weld throats (1 'central' end and 3 'gripped' end)
3	Schenck	0—11	0— 9·4	0·570	Failure from weld end through cover plate
4	Schenck	0—10	0— 8·5	0·734	Failure from end of weld in cover plate.
5	Schenck	0— 9	0— 7·7	1·873	Failure from end of weld through cover plate and crack in weld throat ('gripped' end)
6	Schenck	0— 8·5	0— 7·3	0·918	Failure from weld end through cover plate and 2 cracks in weld throat ('central' ends)
7	Schenck	0— 8·5	0— 7·3	1·293	Failure from weld end in cover plate and crack in weld throat ('gripped' end)
8	Schenck	0— 7·7	0— 6·5	1·542	Failure from weld end in cover plate
9	Los	0— 7·5	0— 6·4	2·227	Small cracks in cover plates at weld ends found in static test. Also transverse crack through weld at crater and crack in weld throat at weld end.
CNO(a)/1	Los	0—16	0—13·7	0·164	Failure from end of weld through cover plate
2	Schenck	0—13·5	0—11·6	0·568	
3	Schenck	0—10·5	0— 9	1·927	
4	Los	0— 7·5	0— 6·4	3·101	Very small cracks at weld ends in cover plate found in static test
CNQ/1	Los	0—13·5	0—11·6	0·066	Failure at weld end in main plate
2	Los	0—10	0— 8·5	0·356	Failure at weld end in main plate and cracks in 3 weld throats
3	Los	0— 6·5	0— 5·6	1·470	Failure at weld end in main plate



5—Fatigue test results for cover plate type specimens failing in the main plates. (Pulsating tension cycle)

The only difference between the two groups forming each pair was the length of the welds, and a comparison of the fatigue strengths obtained at 10^5 and 2×10^6 cycles for the first two pairs is:

Series	*Weld Length, in.*	*Point of Failure*	*Fatigue Strength, tons/sq.in* 10^5	2×10^6	*Theoretical S.C.F. Main Plate*	*Cover Plate*
Stress in cover plate						
CN	4·25	Through cover plates	9·0	3·5	(1·81)	2·07
CN(c)	6·75		9·8	4·4	(1·46)	1·58
Stress in main plate						
CN2	4·25	Through main plate	13·8	4·6	1·91	(1·95)
CN2(a)	6·75		14·9	5·6	1·52	(1·51)

In this table are also shown the theoretical stress concentration factors computed by the method developed by Palmer;[4] it will be seen that these provide at least a qualitative explanation of the greater fatigue strengths obtained with longer welds. A similar trend can be observed when the other two pairs of groups are compared, but both series *CNO(a)* and *CNA(a)* consisted of only 4 specimens so that it is not possible to quote accurate fatigue strengths. In Fig. 6, where groups *CNA* and *CNA(a)* are compared, the open triangles indicate specimens in series *CNA* which failed through the main plate instead of through the cover plate. These points therefore represent a lower limit to the fatigue strength of the cover plates. All the specimens in series *CNA(a)* failed from the weld ends through the cover plates. It can be seen that the approximate fatigue strength at 2×10^6 cycles for

Table VI

Summary of results for series CN2 and CN2(a)

(Specimens with $\frac{1}{2}$ in. thick cover plates, as-welded)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CN2/1	Los	0—15	0—9·65	0·074	Failure on weld throat. Two small cracks in main plate and three in cover plates
2	Los	0—13	0—8·4	0·089	Failure from end of weld through main plate
3	Los	0—12	0—7·7	0·230	Failure at weld ends in cover plates and main plate and also in a weld throat
4	Los	0—10	0—6·4	0·419	Failure at weld ends through main plate
5	Los	0— 7·5	0—4·8	0·727	
6	Schenck	0— 6	0—3·8	1·146	
7	Los	0— 4·5	0—2·9	2·226	Four cracks at weld ends in main plate found in static test
CN2(a)/1	Los	0—14	0—9·0	0·140	Failed from end of weld through main plate
2	Los	0—12	0—7·7	0·272	
3	Los	0—10	0—6·4	0·428	
4	Los	0— 8·5	0—5·5	0·702	
5	Los	0— 6·5	0—4·2	1·680	

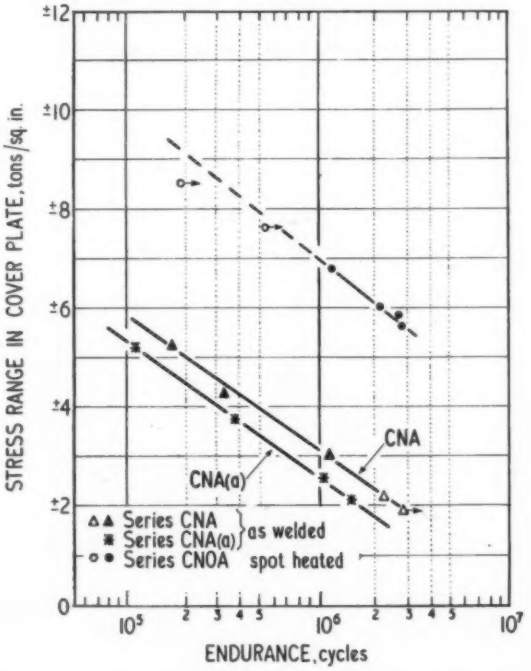
specimens with short welds (series *CNA(a)*) was $\pm 1{\cdot}8$ tons/sq.in. and that this was increased to approximately $\pm 2{\cdot}3$ tons/sq.in. by increasing the weld length, these being the nominal average stresses in the cover plates. In the case of groups *CNO* and *CNO(a)* all the main failures were initiated at the weld ends and propagated through the cover plates. It can be seen from Fig. 4 that at 2×10^6 cycles, series *CNO*, with short welds, gave a fatigue strength of 6·6 tons/sq.in. in the cover plates, but that this was increased to approx. 9 tons/sq.in. for series *CNO(a)*. In drawing the S/N curve for this group the result for specimen *CNO(a)*/4 has been ignored, even though the specimen technically had failed. The fatigue cracks discovered by the static test were extremely small and there is little doubt that, if the specimen had been allowed to run to final rupture, it would have survived for a considerable time, because the rate of crack propagation in spot heated specimens is known to be slow.

To sum up, the effect of increasing weld length was to produce an increase of fatigue strength of approximately 25% for as-welded specimens and 35% for spot heated specimens.

Influence of cover plate thickness

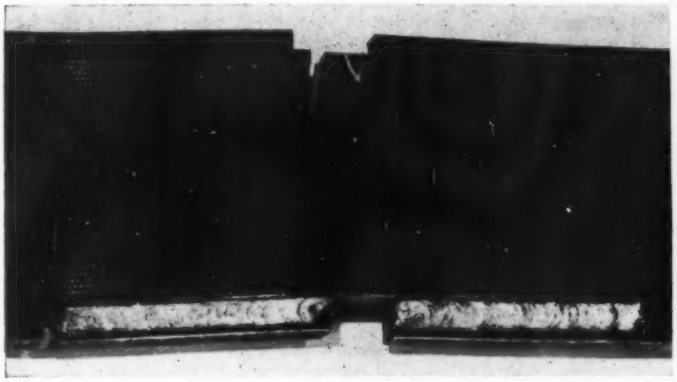
The effect of cover plate thickness on the fatigue strength of as-welded specimens tested in pulsating tension can be seen by comparing the results for series *CN* and *CN2* which were otherwise identical.(Tables IV and VI, Figs. 4 and 5). In series *CN* all the specimens failed through the cover plates, the cracks in specimen *CN*/6 being discovered in the subsequent static test. In series *CN2*, however, all the specimens developed cracks at the ends of the welds in the main plate, although specimens *CN2*/1 and 3 also had cracks in the cover plates. These two results have also been plotted in Fig. 4 and it will be seen that they lie close to the S/N curve for series *CN*. A photograph of the failure in specimen *CN2*/3 is shown in Fig. 7. In specimen *CN2*/7 the fatigue cracks were found in the subsequent static test after the specimen had survived $2{\cdot}2 \times 10^6$ cycles at 0—4·5 tons/sq.in. The fatigue strength at 2×10^6 cycles is shown by these results to be approximately 3·5 tons/sq.in. in the cover plates and 4·6 tons/sq.in. in the main plate. It should be noted that the former figure was confirmed by tests on series *CN2M* which was identical with series *CN2* but was spot heated to prevent failure of the main plates. The results for this series are given in Table VII and have been plotted in Fig. 4.

A similar comparison can be made between series *CNA* and *CN2A*, which were both tested on an alternating cycle. The results obtained are given in Tables VIII and X and are shown graphically in Figs. 6 and 8.



6—Fatigue test results for cover plate type specimens failing in the cover plates. (Alternating cycle)

All the specimens in series *CN2A* failed from the weld ends through the main plate, the failure in specimen *CN2A*/6, tested at ±2 tons/sq.in., being found in the static test after the specimen had survived $4{\cdot}6\times10^6$ load cycles under full load. However, in series *CNA* the position of failure was randomly distributed between the main plate and cover plates and the results therefore have been plotted on both Fig. 6 and



7—*Failure of specimen* CN2/3

Table VII

Summary of results for series CN20, CN20X and CN2M

(Specimens with $\frac{1}{2}$ in. thick cover plates, spot heated)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CN2O/1	Schenck	0—17	0—10·9	0·161	Failed in main plate from weld root approx. $\frac{3}{4}$ in. inside end of weld
2	Los	0—16	0—10·3	0·207	
3	Schenck	0—15	0— 9·6	0·436	
4	Los	0—14	0— 9·0	0·681	
5	Schenck	0—13	0— 8·4	1·031	Failed from end of weld in main plate
6	Schenck	0—12	0— 7·7	2·101	Failed in main plate from weld root approx. $\frac{3}{4}$ in. inside end of weld
7	Schenck	0—11	0— 7·1	1·769	
CN2M/1	Schenck	0—14·5	0— 9·3	0·097	Failed from end of weld in cover plates
2	Schenck	0—11	0— 7·1	0·312	
3	Los	0— 7·5	0— 4·8	0·873	
4	Schenck	0— 5·0	0— 3·2	2·558	
CN2OX/1	Los	0—18	0—11·6	0·091	Failed from ends of welds in main plates
2	Schenck	0—15	0— 9·6	0·334	
3	Schenck	0—12·5	0— 8·0	0·858	
4	Schenck	0—11·5	0— 7·4	1·131	Failed from end of weld in main plate and from surface in heated spot
5	Los	0—10·5	0— 6·7	2·155	Small cracks at weld ends in main plates found in static test

Table VIII

Summary of results for series CNA and CNA(a)

(Specimens with $\frac{3}{8}$ in. thick cover plates, as-welded)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CNA/1	Schenck	±6·0	±5·15	0·170	Failed from end of weld through cover plated
2	Los	±5·0	±4·3	0·312	
3	Schenck	±3·5	±3·0	1·192	Failed from end of weld through main plate. Two small cracks at weld ends in cover plates found ultrasonically
4	Schenck	±3·0	±2·6	1·066	Failed at end of weld through main plate
5	Schenck	±2·5	±2·15	2·142	As for CNA/3
6	Schenck	±2·25	±1·9	2·825	Failed at end of weld through main plate
CNA(a)/1	Schenck	±6·0	±5·15	0·115	Failed from ends of welds through cover plate
2	Los	±4·5	±3·85	0·361	
3	Schenck	±3·0	±2·6	1·054	
4	Schenck	±2·5	±2·1	1·513	

Fig. 8. In Fig. 6, which refers to specimens that failed through the cover plates, the open triangles represent specimens in series *CNA* which failed in the main plate. Thus they form a lower limit to the S/N curve, and the fatigue strength at 2×10^6 cycles can be seen to be approximately $\pm2{\cdot}2$ tons/sq.in. in the cover plates. Figure 8 refers to failure in the main plates; it is shown that there was no significant difference in strength between series *CNA* and *CN2A*, both having a strength at 2×10^6 cycles of $\pm2{\cdot}4$ tons/sq.in. in the main plate.

In the following table a summary of these results, and, for comparison, the theoretical stress concentration factors are given. For specimens with $\frac{1}{2}$ in. thick cover plates the ratio of the cross-sectional areas of main plate: cover plate was 1:1·55 and for $\frac{3}{8}$ in. thick cover plates it was 1:1·17. These figures have been used to compute the peak stresses in the various test series on the basis of unit average stress both in the main plate and in the cover plates. These results are also shown in the table.

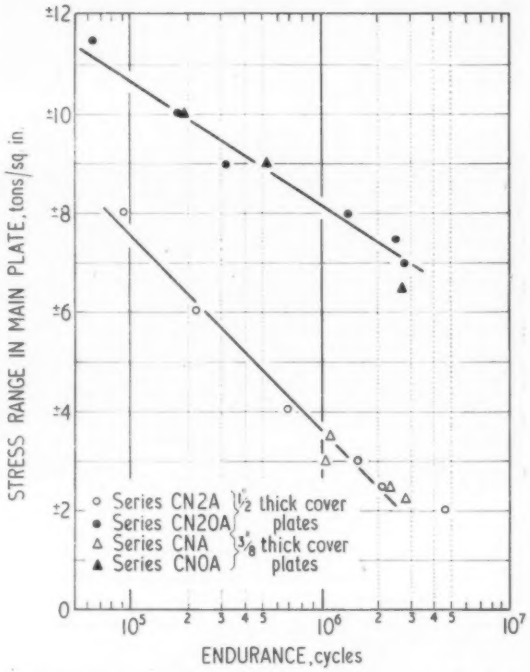
Series	Cover Plate thickness, in.	Fatigue Strength, tons/sq.in. at 2×10^6 cycles		Stress-Concentration factor		Relative Peak Stresses: For Unit Stress in Main Plate		For Unit Stress in Cover Plate	
		Main	Cover	Main	Cover	Main	Cover	Main	Cover
CN	$\frac{3}{8}$	4·6	3·5	1·81	2·07	1·81	1·78	2·12	2·07
CN2	$\frac{1}{2}$		3·5	1·91	1·95	1·91	1·26	2·96	1·95
CN2M	$\frac{1}{2}$								
CNA	$\frac{3}{8}$	$\pm2{\cdot}4$	$\pm2{\cdot}3$	1·46	1·59	1·46	1·36	1·71	1·59
CN2A	$\frac{1}{2}$	$\pm2{\cdot}4$		1·52	1·51	1·52	0·97	2·36	1·51

It is interesting to note that for series *CN* and *CNA* the peak stresses in the main plate and cover plates were approximately equal and that in series *CNA*, but not in series *CN*, this was reflected in the random distribution of failures between the main plates and cover plates. Also in series *CN2* and *CN2A* failures were confined to the main plates in which the peak stresses were about 55% greater than in the cover plates.

Effect of induced residual stresses

Spot heating

The main object of this part of the programme was to compare the fatigue strength of as-welded and spot heated specimens failing in both the main plates and the cover plates under pulsating tension and alternating loads. Eight series of specimens were tested as follows:



8—*Fatigue test results for cover plate type specimens failing in the main plate. (Alternating cycle)*

Series	Cover Plate Thickness, in.	Condition	Load Cycle
CN	$\frac{3}{8}$	As-welded	Pulsating tension
CNO	$\frac{3}{8}$	Spot heated	,, ,,
CN2	$\frac{1}{2}$	As-welded	Pulsating tension
CN2O	$\frac{1}{2}$	Spot heated	,, ,,
CNA	$\frac{3}{8}$	As-welded	Alternating
CNOA	$\frac{3}{8}$	Spot heated	,,
CN2A	$\frac{1}{2}$	As-welded	Alternating
CN2OA	$\frac{1}{2}$	Spot heated	,,

The results for series *CN* and *CNO* are recorded in Tables IV–V, and are shown in the form of S/N curves

Table IX

Summary of results for series CNOA

(Specimens with $\frac{3}{8}$ in. thick cover plates, spot heated)

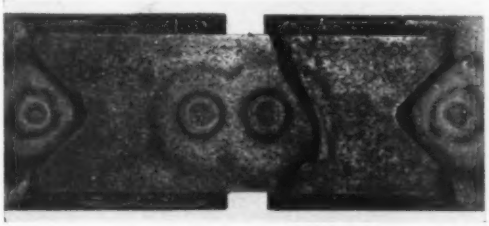
Specimen Ref. No.	Testing Machine	Stress Range, tons/sq.in. Main Plate	Stress Range, tons/sq.in. Cover Plates	Cycles to Failure $\times10^6$	Remarks
CNOA/1	Schenck	±10	±8·55	0·192	Failure from end of weld through main plate
2	Schenck	± 9	±7·7	0·529	
3	Schenck	± 8	±6·85	1·194	Failure from end of weld through cover plates
4	Schenck	± 7	±6·0	2·131	
5	Schenck	±6·75	±5·8	2·648	Failure initiated at 2 weld ends in weld throat and propagated into cover plate $\frac{1}{2}$ in. from weld end
6	Los	± 6·5	±5·6	2·754	Cracks in cover plates and main plate at weld ends found in static test

in Fig. 4. In all the specimens the main failure was initiated at the weld end and the crack propagated through the cover plates. The failure shown in Fig. 9, although not from this group, is typical of those which occurred in series *CN*. In series *CNO* the fractures, although originating at the weld ends, propagated differently from those in series *CN*, tending to skirt around the heat spot. A typical example is shown in Fig. 10. No reasonable explanation has been found for this behaviour. Series *CNO* was also notable for the large number of specimens which developed fatigue cracks in the weld throats, although these were never the primary cause of failure (Fig. 11). This behaviour will be considered in more detail later. It will be seen from Fig. 4 that at 2×10^6 cycles the effect of spot heating was to increase the fatigue strength of the cover plates from approximately 3·5 to 6·5 tons/sq.in.

The results for series *CN2* and *CN2O* are given in Tables VI–VII, and are plotted in Fig. 5. Since these specimens had $\frac{1}{2}$ in. thick cover plates it was hoped that failures would occur consistently in the main plates. This was the case with series *CN2O*, but in the as-welded series (*CN2*) two specimens also failed through the cover plates. One of these, *CN2/3*, is illustrated in Fig. 7. In series *CN2O* only one specimen failed from the extremity of the weld. The remainder failed about $\frac{3}{4}$ in. inside the weld, the actual site of initiation being the weld root; a typical failure of this type is illustrated in Figs. 12 and 13. The change in position of failure must be ascribed to the effect of spot heating. The fatigue strength of the main plates can be seen from Fig. 5 to be increased from about 4·7 to 11·5 tons/sq.in. by spot heating.

In the tests on an alternating load cycle it was hoped that series *CNA* and *CNOA*, each having $\frac{3}{8}$ in. thick cover plates, would fail through the cover plates, and that series *CN2A* and *CN2OA* would fail through the main plates. However, as can be seen from Tables VIII–XI this was only partially the case. All the specimens with $\frac{1}{2}$ in. thick cover plates did fail through the main plates, but so did many of the specimens in series *CNA* and *CNOA*. The results for those specimens which failed through the cover plates are plotted in Fig. 6, in which the 'open' points represent specimens which failed in the main plates and which therefore form a lower limit to the S/N curves. In view of the small number of definite results it is difficult to state accurate fatigue strength values, but it can be seen that the effect of spot heating was to increase the strength at 2×10^6 cycles from approximately $\pm2{\cdot}3$ to ±6 tons/sq.in. in the cover plates.

9—*Typical failure through cover plates from weld ends*



10—*Failure of specimen* CNO/5

11—*Cracks in weld in specimen* CNO/9

12—*Failure of specimen* CN2O/6

13—*Fracture surface of specimen* CN2O/6 *showing failure from the weld root*

Table X

Summary of results for series CN2A

(Specimens with ½ in. thick cover plates, as-welded)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CN2A/1	Schenck	±8	±5·15	0·094	Failure from weld end in main plate
2	Schenck	±6	±3·85	0·222	
3	Schenck	±4	±2·57	0·693	
4	Los	±3	±1·93	1·617	
5	Schenck	±2·5	±1·60	2·110	
6	Schenck	±2	±1·29	4·631	Small crack in main plate at weld end found in static test

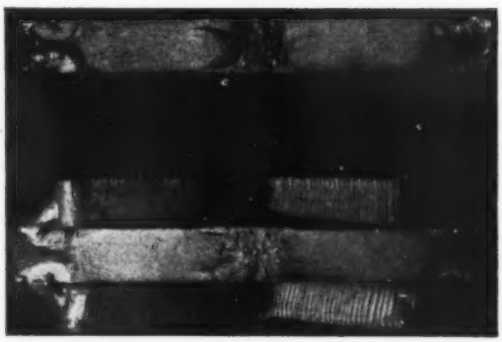
The results for specimens which failed through the main plates are plotted in Fig. 8; it will be seen that there was no significant difference in strength between series *CNA* and *CN2A* and between *CNOA* and *CN2OA*. At 2×10^6 cycles the effect of spot heating was to increase fatigue strength from ± 2·5 to ±7·5 tons/sq.in. As in series *CN2O*, all except two of the specimens in series *CN2OA* failed inside the end of the weld from the weld root. In specimens *CN2OA*/4, 5, and 6 the cover plate also failed from the same place. Five specimens, *CN2OA*/2, 4*x*, 6*x*, 7*x*, and 8*x* which failed from the heated spot in the main plate, were found to have been overheated. A typical macro-section of such a heated spot is illustrated in Fig. 2*b*. These five specimens exhibited two typical modes of failure, one being through the centre of the heated spot initiated at 'porosity' in the centre of the plate (Figs. 14 and 2*b*), the second being at the edge of the spot initiated at the surface where there was a discontinuity in thickness associated with a large grain structure (Figs. 15 and 2*b*).

Some subsidiary tests were carried out also on specimens spot heated by oxyacetylene torches. One of these series, *CN2OX*, is directly comparable with series *CN2O* (Table VII, Fig. 5). It is apparent that heating with oxyacetylene torches was not quite so effective as heating in the spot welding machine, the fatigue strength at 2×10^6 cycles dropping by about 1 ton/sq.in. to 10·5 tons/sq.in.

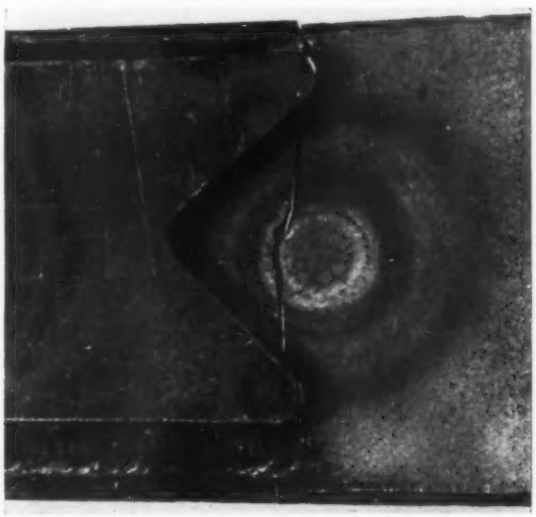
In series *CN2M* and *CNQ* only the main plates and cover plates respectively were heated. This resulted in the transference of the point of failure to the cover plates for series *CN2M* and to the main plates for series *CNQ*. Thus series *CN2M* became equivalent to series *CN* and it can be seen from Fig. 4 that the results for these groups were virtually identical. Insufficient specimens were tested in series *CNQ* to define fatigue strength but the results are shown in Table V and Fig. 5. The main object in this case, to determine whether the point of failure could be moved, was achieved.

Local compression

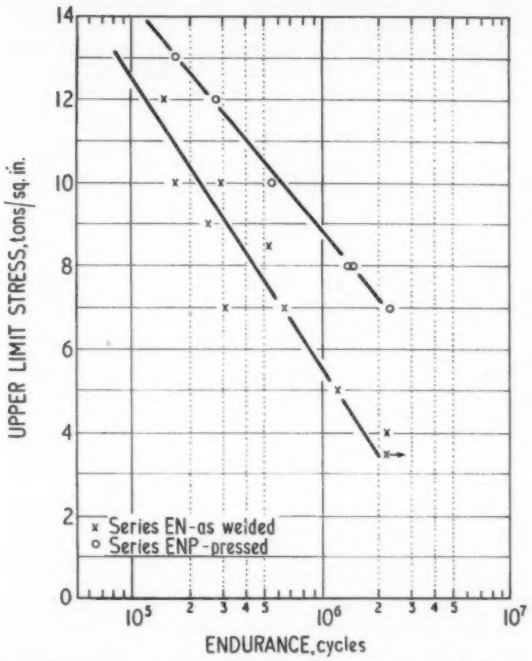
Four series of specimens were tested to determine the effect of local compression on both a pulsating tension and an alternating load cycle. The results for series *EN* and *ENP*, tested in pulsating tension, are set out in Tables XII and XIII and are shown in the form of S/N curves in Fig. 16. The majority of the specimens in these two series failed from the weld ends in either the gripped or central plates, but in specimen *ENP*/1, tested at 13 tons/sq.in. the primary failure was in the weld throat. Specimen *EN*/2, tested at 10 tons/sq.in., also had a small crack in the weld throat. In specimen *EN*/9, tested at 4 tons/sq.in. for $2{\cdot}2\times10^6$ cycles, the fatigue cracks at the weld ends were only found in the subsequent static test, and in specimen *EN*/10, which survived $2{\cdot}2\times10^6$ cycles at



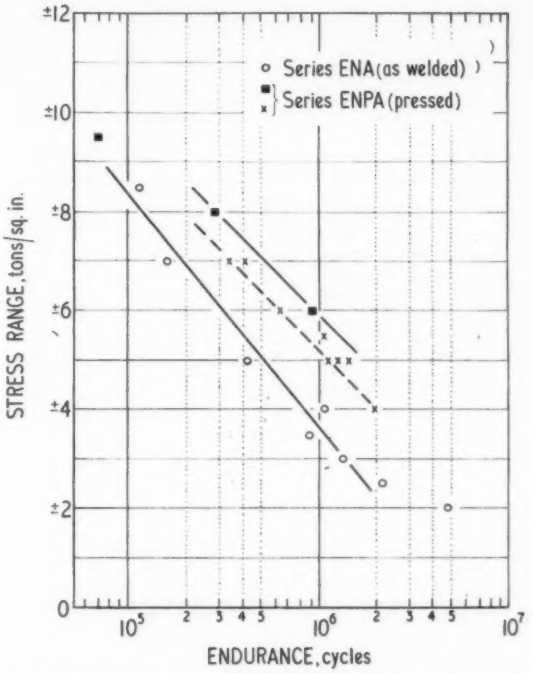
14—*Failure from 'porosity' at centre of overheated spot in specimen* CN2OA/4x



15—*Failure from surface at edge of overheated spot in specimen* CN2OA/6x



16—*Fatigue test results for egg box type specimens.* (*Pulsating tension cycle*)



17—*Fatigue test results for egg box type specimens.* (*Alternating cycle*)

3·5 tons/sq.in., no cracks were found in the subsequent static test. It can be seen from Fig. 16 that the effect of the local compression was to produce a 100% increase in strength at 2×10^6 cycles, from 3·5 to 7·0 tons/sq.in.

Series *ENA* and *ENPA* were tested under an alternating cycle and the results are recorded in Tables XIV and XV and in Fig. 17. In series *ENA* all the specimens failed from the weld ends, but in series *ENPA* the majority of specimens failed from the slot at the mid length of the welds and only three valid results were obtained for direct comparison with series *ENA*. However, all the results have been plotted in Fig. 17, the crosses representing those specimens which failed from the slot and which can be regarded therefore as forming a lower limit to the S/N curve for series *ENPA*. Hence the fatigue strength at 2×10^6 cycles was increased from about $\pm$ 2·5 tons/sq.in. to at least $\pm$ 4 tons/sq.in. The change in the point of failure presumably occurred because the local compression treatment blanketed the notch at the weld ends.

Failures in the welds

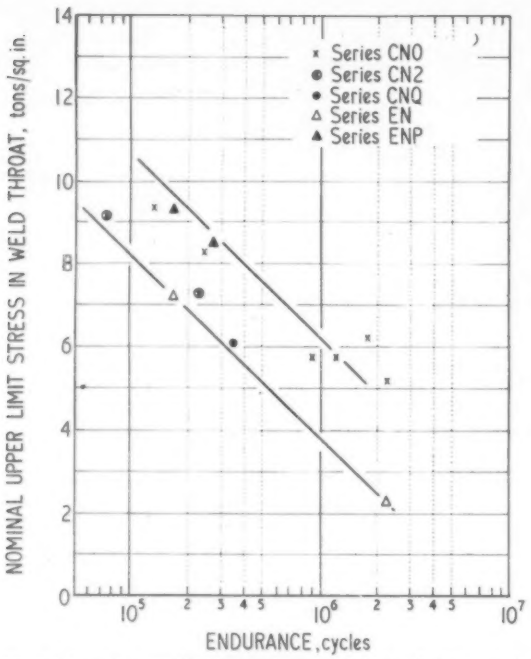
As we stated earlier, during the course of the investigation several failures occurred in the welds and a summary of these is given in Table XVI. It can be seen that all the specimens exhibiting this type of failure had short welds and that many of them were either spot heated or locally pressed. Both these facts

Table XI

Summary of results for series CN2OA

(Specimens with ½ in. thick cover plates, spot heated)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in. Main Plate*	*Cover Plates*	*Cycles to Failure* $\times 10^6$	*Remarks*
CN2OA/1	Los	$\pm$11·5	$\pm$7·4	0·064	Failure from end of weld through main plate
2	Schenck	$\pm$10	$\pm$6·45	0·189	Failed from heat spot and from ½ in. inside weld end through main plate
3	Schenck	$\pm$ 9	$\pm$5·8	0·310	Failure from weld ends through main plate
4	Schenck	$\pm$ 8	$\pm$5·15	1·449	Failed from root of weld 1 in. from weld end through main plate. Cover plate failed from same point
5	Los	$\pm$ 7·5	$\pm$4·8	2·575	
6	Schenck	$\pm$ 7	$\pm$4·5	2·818	
4x	Schenck	$\pm$ 8		0·493	Failed from heated spot in main plate
6x	Schenck	$\pm$ 7		0·576	
7x	Schenck	$\pm$ 6·5		1·306	
8x	Schenck	$\pm$ 6		0·927	



18—*Fatigue test results for specimens exhibiting fatigue cracks in the weld. (Pulsating tension cycle)*

imply that the specimens were tested at fairly high stresses, but in fact the stresses on the nominal throat areas were not particularly high. This can be seen from Fig. 18 where the fatigue test results for specimens with cracked welds have been plotted.

In two specimens, *ENP*/1 and *CN*2/1, the failure on the weld throat was the main failure, but in all other cases the cracks in the welds formed subsidiary failures, the main failures being through the main plates or cover plates as discussed previously. The great majority of weld cracks were initiated at the weld ends at the point where the plane of the weld throat first penetrated the parent material. At this point there is a stress concentration which can be shown theoretically to be at least 1·5–2 for the cases under consideration, and no doubt this accounts for the failures. Three of the specimens, however, showed very interesting forms of failure. In specimens *ENP*/2 and *CNO*/9 cracks occurred through the weld craters, which were within the weld length, transverse to the direction of the weld, and in specimen *EN*/9 a crack occurred in the plane of the throat through the weld crater, which was again within the weld length. The cracks occurring in specimen *CNO*/9 are shown in Fig. 11.

From these results it would appear that the fatigue strength of load-carrying welds as specified in design standards may be too high if a weld end, and hence a stress concentration, is involved. In addition, the fatigue strength of load carrying welds, even without the influence of weld ends, may be seriously diminished by crater defects.

Table XII

Summary of results for series EN

(Egg box type specimens, as-welded)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in.*	*Cycles to Failure ×10⁶*	*Remarks*
EN/1	Los	0—12	0·153	In all specimens failure occurred at weld ends, either in gripped plates or central plate. *EN*/2 also had small crack in weld throat beginning of weld run (gripped end).
2	Los	0—10	0·173	
3	Los	0—10	0·295	
4	Los	0— 9	0·253	
5	Los	0— 8·5	0·534	
6	Los	0— 7	0·308	
7	Los	0— 7	0·649	
8	Los	0— 5	1·207	
9	Los	0— 4	2·215	Cracks at weld ends found in static test. Also cracks in weld throat at crater and at weld end.
10	Los	0— 3·5	2·203	No cracks found in static test.

Conclusions

From the experimental work carried out the following conclusions can be drawn:

(1) The fatigue strength of as-welded mild steel specimens of the 'cover plate' type containing longitudinal load-carrying fillet welds was found to lie between approx. 3·5 and 5·5 tons/sq.in. at 2×10^6 cycles, and between 9 and 15 tons/sq.in. at 10^5 cycles,

Table XIII

Summary of results for series ENP

(Egg box type specimens, locally pressed)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in.*	*Cycles to Failure ×10⁶*	*Remarks*
ENP/1	Schenck	0—13	0·177	Failure in weld throat and small cracks at ends of 2 welds
2	Los	0—12	0·279	Failure from weld ends. Also crack in weld throat and crack transverse to weld length through crater.
3	Los	0—10	0·555	Failure at weld ends either in gripped or central plates.
4	Schenck	0— 8	1·493	
5	Schenck	0— 8	1·501	
6	Schenck	0— 7	2·264	

Table XIV

Summary of results for series ENA

(Egg box type specimens, as-welded)

Specimen Ref. No.	*Testing Machine*	*Stress Range, tons/sq.in.*	*Cycles to Failure ×10⁶*	*Remarks*
ENA/1	Los	±8·5	0·119	All failures at weld ends in centre plate, main plates or both together.
2	Schenck	±7	0·165	
3	Schenck	±5	0·422	
4	Los	±4	1·096	
5	Schenck	±3·5	0·873	
6	Los	±3	1·377	
7	Los	±2·5	2·163	
8	Schenck	±2	4·9	Fatigue crack at weld end in centre plate in static test.

Table XVI

Summary of results for specimens which cracked in the welds

Specimen Ref. No.	Stress Range, tons/sq.in. Main Plate	Weld Throat	Cycles to Failure $\times 10^6$	Remarks
ENP/1	0—13	0—9·28	0·177	Failure on weld throat (main failure)
CNO/1	0—13·5	0—9·25	0·147	Crack on weld throat (subsidiary failure)
CN2/1	0—15	0—9·06	0·074	Failure on weld throat (main failure)
ENP/2	0—12	0—8·56	0·279	Crack transverse to weld direction through crater. Also crack in throat
CNO/2	0—12	0—8·24	0·245	Failure on throat at one or more weld ends
CN2/3	0—12	0—7·25	0·230	
EN/2	0—10	0—7·15	0·173	
CNO/5	0— 9	0—6·18	1·873	
CNQ/2	0—10	0—6·05	0·356	
CNO/6	0— 8·5	0—5·84	0·918	
CNO/7	0— 8·5	0—5·84	1·293	
CNO/9	0— 7·5	0—5·15	2·227	Transverse crack through crater and crack on throat at end
EN/9	0— 4	0—2·29	2·214	Longitudinal crack through crater and crack at weld end

for a pulsating tension load cycle. For an alternating load cycle the strength range was about ± 6 to $\pm 7{\cdot}5$ tons/sq.in. at 10^5 cycles and $\pm 2{\cdot}0$ to $\pm 2{\cdot}5$ tons/sq.in. at 2×10^6 cycles. The lower limits refer to stresses in the cover plates and the upper limits to stresses in the main plates, intermediate values being associated with either. The strengths are summarised as follows:

Cover Plate Thickness, in.	Weld Length, in.	Fatigue Strength, tons/sq.in. Strength of Main Plate Series	10^5	2×10^6	Strength of Cover Plate Series	10^5	2×10^6
Pulsating tension cycle							
¾	4·25	—	—	—	*CN*	9·0	3·5
	6·75	—	—	—	*CN(c)*	9·8	4·4
½	4·25	*CN2*	13·8	4·6	*CN2M*	9·0	3·5
	6·75	*CN2(a)*	14·9	5·6	—	—	—
Alternating cycle							
¾	3·75	—	—	—	*CNA(a)*	$\pm 5{\cdot}3$	$\pm 1{\cdot}8$
	6·75	—	—	—	*CNA*	$\pm 5{\cdot}8$	$\pm 2{\cdot}3$
½ & ¾	6·75	*CNA*, *CN2A*	$\pm$ 7·6	$\pm 2{\cdot}5$	—	—	—

(2) For the particular geometry selected, it was found that, in general, cover plate type specimens with ¾ in. thick cover plates (ratio of cross-sectional area of main plate: area of cover plate = 1:1·17) failed through the cover plates, but that specimens with ½ in. thick cover plates (area ratio = 1:1·55) failed through the main plates. There were, however, some exceptions to this rule.

(3) Weld length was found to be an important variable in determining fatigue strength, greater length resulting in enhanced fatigue strength of both the main plates and cover plates, as can be seen from the Table above. Qualitatively this behaviour can be explained by the fact that the theoretical stress concentration factors are reduced by increasing the weld length.

Table XV

Summary of results for series ENPA

(Egg box type specimens, locally pressed)

Specimen Ref. No.	Testing Machine	Stress Range, tons/sq.in.	Cycles to Failure $\times 10^6$	Remarks
ENPA/1	Schenck	$\pm 9{\cdot}5$	0·068	Failed from weld end in centre plate
2	Schenck	± 8	0·287	
3	Los	± 6	0·915	Failed from slot and from weld end in centre plate
1x	Schenck	± 7	0·337	Failed from end of slot
2x	Schenck	± 7	0·406	
3x	Schenck	± 6	0·647	
4x	Los	$\pm 5{\cdot}5$	1·025	
5x	Schenck	± 5	1·146	
6x	Schenck	± 5	1·250	
7x	Schenck	± 5	1·455	
8x	Schenck	± 4	1·959	

(4) The effect of spot heating was to increase the fatigue strength at 2×10^6 cycles by 88% to 200%. The results obtained are summarised as follows:

	Fatigue Strength, tons/sq.in. Pulsating Tension 10^5	Pulsating Tension 2×10^6	Alternating 10^5	Alternating 2×10^6
Specimen Condition				
Failure in the main plate				
As-welded	13·8	4·6	$\pm$ 7·5	$\pm 2{\cdot}5$
Spot heated	18·0	11·5	$\pm 10{\cdot}7$	$\pm 7{\cdot}5$
Increase of strength, %	30	150	42	200
Failure in cover plates				
As-welded	9·0	3·5	$\pm$ 5·8	$\pm 2{\cdot}3$
Spot heated	12·1	6·6	—	$\pm 6{\cdot}0$
Increase of strength, %	33	88	—	160

(5) The effect of local compression was to increase fatigue strength at 2×10^6 cycles by about 100% for both pulsating tension and alternating loading. The results are summarised as follows:

	Fatigue Strength, tons/sq.in. Pulsating Tension 10^5	Pulsating Tension 2×10^6	Alternating 10^5	Alternating 2×10^6
Specimen Condition				
As-welded	12·5	3·5	$\pm$ 8·4	$\pm 2{\cdot}2$
Locally pressed	14·5	7·0	—	$\pm 4{\cdot}5$
Increase of strength, %	16	100	—	100

Acknowledgments

The author wishes particularly to acknowledge the help and encouragement he has received from Mr. R. P. Newman during the progress of this work. He is also grateful for the help given in the preparation and testing of the specimens by his colleagues, Messrs. H. J. Tuck, E. O. Lawrance, and R. Males.

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3. T. R. GURNEY: *Brit. Welding J.*, 1960, vol. 7, pp. 415–431.
4. P. J. PALMER: *Brit. Welding J.*, 1955, vol. 2, pp. 55–60.

NEWS and Announcements

News of the Institute and Branches
B.W.R.A. · Other Societies ·
and Industry

INSTITUTE ACTIVITIES

MEETINGS

Christmas Lecture

The Annual Lecture to Young People will be given at the Institute on Thursday, 4th January 1962. The lecturer will be Mr. John Strong, and tickets will be available by the beginning of December.

Members are advised to make early application on behalf of young members of their families.

Fifth Annual Lecture

Dr. P. F. R. Venables, Principal of the City of Birmingham College of Advanced Technology has accepted the President's invitation to give the Fifth Annual Lecture. His subject will be "*The Place of Welding in the National Programme of Technical Education*". The date of the lecture will be 22nd March 1962 (not 8th March) as announced in the London Meetings Programme.

WELDING ENGINEERING 1962

Nearly forty companies have already booked space for Welding Engineering 1962, the Institute of Welding Exhibition in connection with the Spring Meeting at Buxton next May. This will be the most representative display of British welding equipment to be held in connection with an Institute meeting, and a very large attendance is expected.

PHYSICS OF ARC WELDING

Some eighteen of the authors who will be presenting thirty papers on different aspects of the Physics of the Welding Arc at the Autumn Meeting next year, met at the Institute on 26th September last under the chairmanship of Mr. J. F. Lancaster to discuss the planning of the symposium. Among those present were:

Mr. J. C. Amson (United Kingdom)
Dr. O. Becken (Germany)
Dr. M. Bertung (Denmark)
M. F. Dennery (France)
M. L. F. Defize (Netherlands)
Mr. K. Goldman (United Kingdom)
Mr. L. A. King (United Kingdom)
Dr. Mantel (Germany)
Dr. D. R. Milner (United Kingdom)
Dr. J. C. Needham (United Kingdom)
M. A. Rudaz (Switzerland)
Dr. G. R. Salter (United Kingdom)
M. J. F. Sunnen (Belgium)
Mr. A. A. Smith (United Kingdom)

The papers will fall into four groups dealing respectively with Arc Structure, Electrical Characteristics, Metal Transfer and Metallurgy.

In the afternoon Monsieur A. Leroy, Scientific and Technical Secretary of the International Institute of Welding and Mr. G. Parsloe, Secretary General, presided jointly over a session of the same meeting at which Monsieur Leroy explained the proposals for the formation of a Study Group on Welding Arc Physics within the International Institute of Welding and sounded the opinion of the experts present. There was unanimous support for the proposal to set up a small permanent Study Group on an international basis and to empower it from time to time to convene larger gatherings of workers in this field, similar to the 1962 Autumn Meeting of the British Institute.

ELECTION OF MEMBERS

The following elections have been made effective from the dates indicated:

Members

H. B. Cary (Ohio, U.S.A.); *C. E. Jackson (New Jersey, U.S.A.)—25.7.61.

Associate Members

F. C. Kay (Australia) —23.2.61
A. R. K. Bawden (Lincs.); H. Horton (Lancs.); G. I. Lewis (Hants.)—25.7.61

Companions

R. Bennett (Herts.); C. T. Marsh (Lancs.) —25.7.61

Graduates

S. K. Sarkar (India) —15.12.60
D. S. Crighton (Kent); P. K. Ghosh (India) —15.6.61
A. D. Daghlian (Iraq); R. E. Fretwell (Somerset) —25.7.61

Associates

R. Hill (Glasgow); D. L. Thompson (Essex); K. Wootton (Warwicks.) —15.6.61
A. R. Cuthbertson (Co. Durham); W. G. Goldfinch (London); B. M. O. Jones (London); J. McDougall (Co. Durham); †S. P. Rao (India); R. Stewart (Mddx.); G. D. Stitt (Bristol) —25.7.61

*Transfer from Associate Member.
†Transfer from Graduate.

NEWS OF MEMBERS

Mr. G. W. Lanfear has been appointed Welding Engineer to the Decca Group of Companies.

Mr. L. F. Denaro has relinquished his position as Assistant to the Managing Director of the International Nickel Co. (Mond) Ltd. to become General Manager of Publicity, in succession to Mr. A. C. Sturney.

Mr. R. Butler, formerly a Director of Quasi-Arc Ltd. (B.O.C.), has been appointed Managing Director of Eutectic Welding Alloys Co. Ltd.

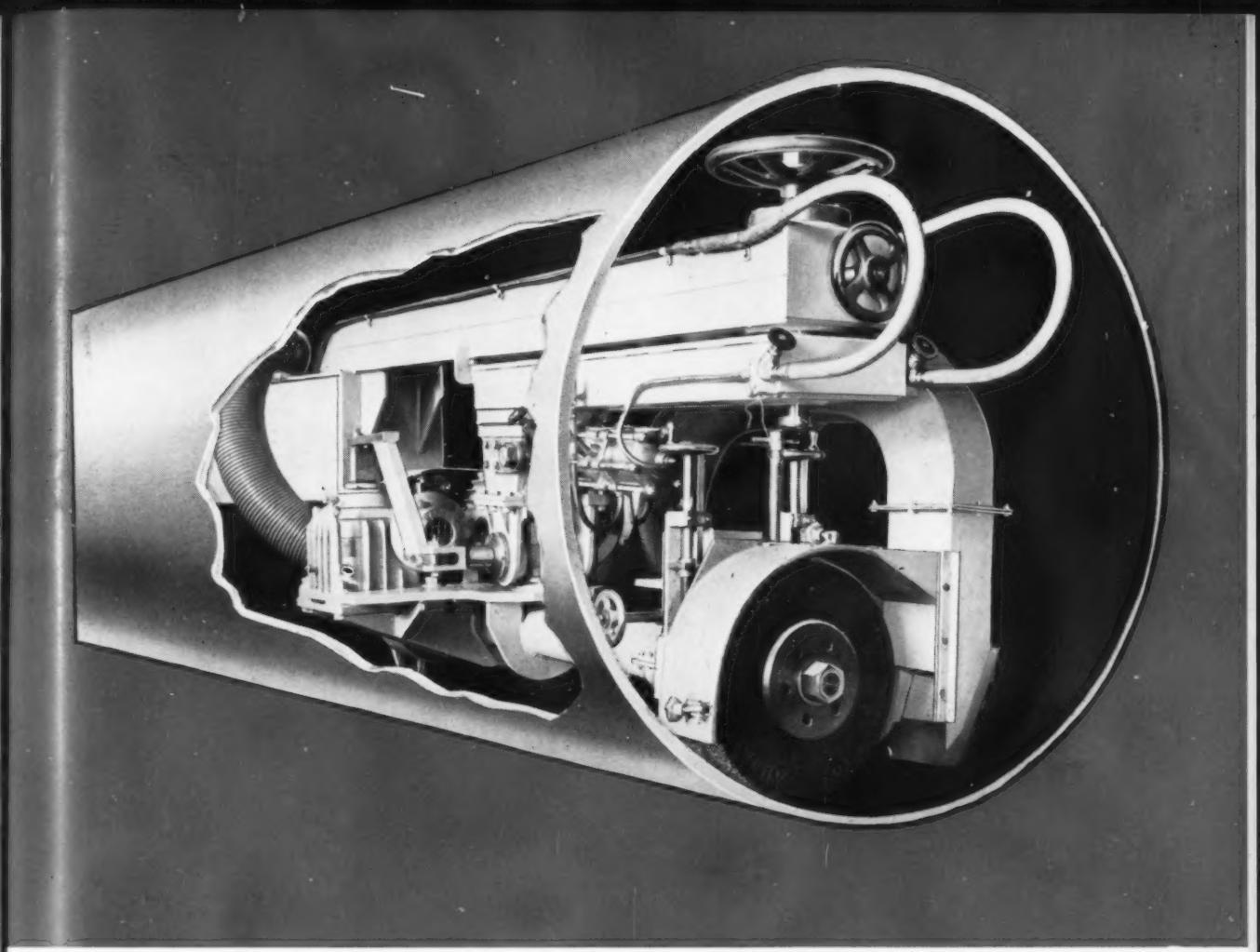
Obituary

The Council regrets to record the deaths of the following members:

Mr. C. Fleet (South London, Member 1923)

Mr. N. F. Wheeler (Southern Counties, Assoc. Member 1945).

Mr. R. E. Blackburn (North London, Slough, Assoc. Member 1937).



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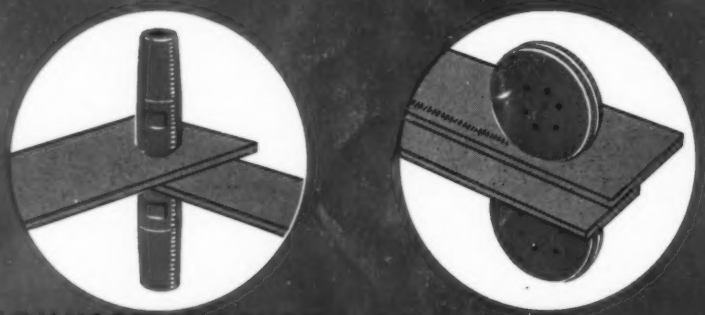

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CONTRIBUTORS TO THE JOURNAL

R. D. Johnston, B.Sc., A.I.M., who was born in 1930, is in the Department of Metallurgy at Battersea College of Technology. He was educated at Bedford School after which he entered the College (then Battersea Polytechnic) where he took Intermediate B.Sc. and an

Honours Degree in Metallurgy as an internal student of the University of London. During this time, he also obtained an Associateship of the Institution of Metallurgists, and in 1957 was awarded a British Oxygen Research Fellowship, enabling him to carry out work on the welding of rare metals, in conjunction with the U.K.A.E.A.

In September 1960 Mr. Johnston joined the teaching staff at Battersea, and is at present engaged on further research as a development of the work outlined in his paper, published in this issue of *BWJ*.

BRANCH NEWS

Sessional Programmes

N.E. Tees-side

The meeting announced for 18th December (*BWJ*, October, p.504) has been cancelled. The revised date and lecture will be given later in the *Diary*.

N.E. Tyneside

The title of the lecture by Dr. Tylecote on the 4th January (*BWJ*, September, p.460) is '*The Archaeology of Metal Fabrication*" and not as published.

South Western

The dates of the sessional meetings announced in *BWJ*, September, p.461) are as follows: 5th Oct.; 6th Nov.; 5th Dec.; 9th Jan.; 5th Feb.; 5th March.

In addition, the Annual Dinner will be held at the Grand Hotel, Broad Street, Bristol, on 8th December, and the Annual General Meeting will be held at Radiant House, Bristol, on 9th April.

Meetings commence at 7.15 p.m., except that on the 5th December which will start at 6.15 p.m. and will be held at the Congregational Hall, Avonmouth.

North London **Slough Section**

3rd Oct.—"*Choice of welding materials*" by L. M. Gourd.
15th Dec.—"*The all-welded Maidenhead Bridge*" by J. S. Allen
6th Feb.—Demonstration of fine wire welding—Government Training Centre, Slough
23rd March—Joint Annual Dinner—Quaglino's Restaurant
3rd April—Annual General Meeting, followed by Question Evening, "*Any welding queries?*"

Meetings are held in the Lecture Room, Slough Community Centre, Farnham Road, commencing at 7.30 p.m.

South London

11th Oct.—"*Metallurgical aspects of resistance welding*" by D. R. Thorneycroft
8th Nov.—"*Fatigue and its effect on welding design*" by T. R. Gurney
13th Dec.—"*Modern welding techniques and inspection of aluminium welds*" by D. G. W. Claydon and H. C. Constantine. (Joint Meeting with North London Branch)
10th Jan.—"*Welding for low-temperature service*" by R. E. Lismer
14th Feb.—"*Welding in the oil industry*" by G. W. Benzie
14th March—"*The use of welding on London Transport Railways*" by E. E. Laird
23rd March—Joint Annual Dinner—Quaglino's Restaurant
11th April—Annual General Meeting followed by Continental Films

Meetings are held at 54 Princes Gate, commencing at 7.30 p.m.

Medway Section

27th Sept.—"*Economics of arc welding*" by J. K. Johannesen
1st Nov.—"*Terminal harbours and transcontinental pipeline*" by N. W. Grey—at Fred. Braby & Co. Ltd., Crayford
7th Feb.—"*Fabrication of heat exchangers for Bradwell Power Station*" by T. E. M. Jones and B. K. Twigg
7th March—Demonstrations of automatic and semi-automatic equipments using inert-gas processes—Medway College of Technology, Chatham
23rd March—Joint Annual Dinner—Quaglino's Restaurant
16th April—Annual General Meeting

Except where otherwise indicated, meetings are held at The Sun Hotel, Chatham, commencing 7.30 p.m.

Branch Committees 1961–62

The following honorary officers and members of committees have been appointed for the session 1961–62.

Indian

President	SIR L. P. MISRA
Chairman	B. N. BANERJEE
Vice-Chairman	S. K. PATHAK
Hon. Treasurer	B. N. BHATTACHARJEE
Hon. Secretary	B. N. MAJUMDAR

Committee

S. BASU	P. K. MALLICK
R. GHOSH	S. S. MUKHERJEE
P. C. KAVANAGH	J. A. MULIYIL
	V. VENKATESAN

BRITISH WELDING RESEARCH ASSOCIATION

"*Automatic Welding*"

Translation into English of all twelve issues of the Russian welding journal '*Avtomaticheskaya Svarka*' for 1960 has now been completed by the Association.

A particular feature of this journal, the translation of which is sponsored by the Department of Scientific and Industrial Research, is its 'broad spectrum' readability. Science, in the Soviet Union, is regarded very much as an industrial tool and most of the Institutes concerned with research are also the designers of the equipment and machinery used for its application to production. As a result, the relationship between research and its practical aspects – how it is related to improving quality, increasing production or improving the lot of the operator—tends to be made much more clear than is often the case in Britain. Vigorous and critical appraisals of techniques and equipment by personnel at all levels are regularly published in the appropriate journals.

Considerable interest was aroused in the West by earlier reports appearing in *Automatic Welding* of the development and extensive industrial application in Russia of electroslag and friction welding. The former technique has now been applied to heavy constructional engineering in the U.K. and the latter, following the building of a prototype friction welder by B.W.R.A., will shortly take its place among other production welding processes in the U.K.

Already, as the 1960 issues of *Automatic Welding* show, both techniques

have found further application; electroslag welding as a means of remelting and refining metals; friction welding as a hardfacing and rebuilding technique. Further, with the use of rotated metal inserts, friction welding is now applied to a much wider range of products than has hitherto been possible.

One of the principal aims in the Russian engineering industry is to mechanise processes wherever this can be done, and the extensive use of prefabrication methods, even with such structures as large-capacity oil tanks and blast-furnace casings, enables much of the work on these structures to be carried out by automatic and semi-automatic means. Rolled structural members are, to an increasing extent, being replaced by welded components built up from coiled sheet steel and, during 1960, a national competition, with cash prizes of up to 10,000 roubles, was organised with the object of further increasing mechanisation of welding processes and of replacing orthodox methods of manufacture (casting, rolling, forging) by welded fabrication.

A technique first described in *Automatic Welding* in 1959, and at that time regarded in Britain largely as a curiosity —the use of steam as a shielding medium —has been further exploited and evidence is given that the process is being used on a production scale in several Russian engineering works. In common with friction welding the process was considered initially as a means of overcoming the difficulties of power supply to outlying sites but its use has been extended to works where such problems do not arise.

While much of the Russian work parallels that carried out in Britain, the regular appearance in *Automatic Welding* of papers on non-conventional, but often surprisingly practical, techniques, coupled with forthright discussions at all levels on all aspects of welding and mechanisation, fully justifies the decision of D.S.I.R. to sponsor its translation.

INTERNATIONAL INSTITUTE OF WELDING

1961 Annual Assembly in New York

A brief report on this Assembly was given in "*International Welding*", No. II (*BWJ*, Supplement, June 1961, p. 36*s*). The following is a more complete report issued by the IIW.

For the first time since its foundation in 1948, the IIW held its Annual Assembly in 1961 outside Europe. The American Welding Society (AWS), on behalf of the American and Canadian Councils of the IIW, organised this Assembly in New York from 11th to 18th April, 1961. Some 530 persons (delegates, experts and observers and those accompanying them) from twenty-two member countries attended and were invited to be present also at the forty-second Annual Meeting of the AWS held in the following week. Many tributes were paid to the Organising Committee and the AWS staff for the outstandingly successful organisation of these two large meetings in consecutive weeks.

Programme

After preliminary meetings of the Executive and Governing Councils, the Assembly was formally opened on 12th April at a session at which Mr. W. Edström, President of the IIW, Mr. R. D. Thomas Jr., President of the American Welding Society, and Mr. A. B. Kinzel, member of the National Academy of Sciences, spoke.

From 12th to 15th April each of the fifteen technical Commissions of the Institute held four meetings which constituted the principal business of the Assembly. Particulars of some of the work accomplished and of decisions taken by the Commissions are given below.

At the end of the Commissions' sessions and before the presentation of their resolutions to the Governing Council, delegates and experts held meetings by countries to consider, from their respective national viewpoints, the results of the Commissions' work.

The Governing Council considered the Commissions' resolutions at a meeting on 18th April which was followed by a meeting of the Executive Council.

Decisions of the Governing Council

(*a*) *Vice-Presidency of the IIW*

The terms of office as Vice-President of the IIW of Professor H. Kihara (Japan) having expired this year, Professor J. Çabelka (Czechoslovakia) was elected to succeed him.

(*b*) *New Member Country*

The Israel Institute of Metals was elected to membership. This is the first Israeli body to join the IIW and its election brings the number of member countries to twenty-nine.

(*c*) *Constitution of the IIW*

The Constitution of the IIW was amended so that the Chairmen of Commissions are appointed, not for an indefinite period, as hitherto, but for a period of five years with the possibility of immediate reappointment. As a result of this amendment, with effect from 1962, the terms of office of the Chairmen of three of the fifteen Commissions, taken in sequence, will end each year.

Work of the Commissions

(*a*) *Documents for Publication*

As usual, member societies of the IIW will have the opportunity to publish, in their own journals and in the specialised reviews of their countries, the reports prepared by the Chairmen of the Institute's Commissions on the work carried out since the Assembly in Liège (Belgium) in 1960 and during the New York meetings.

In addition, various documents, some presented to the Commissions by their authors, others prepared by the Commissions themselves, one which is a summary of a document of which the complete text is available but not altogether suitable for publication, and finally one emanating from the Scientific and Technical Secretariat, were recommended for publication in the reviews of member countries. These documents, which number seventeen, are as follows:

- The present state of development of the applications of gas pressure welding in certain member countries of the IIW.
- Comparative study of butt-welding methods of steels and non-ferrous metals used in different countries.
- Considerations on the influence of various factors in the magnetic particle inspection of surface defects in welds.
- Report on the use of the controlled thermal severity (CIS) test in different laboratories.
- Calculation and measurement of residual stresses due to spot heating.
- Investigation on the shrinkage due to multiple spot heating.
- Radiography of welds on boilers and pressure vessels (First Part: Principles).
- Porosity in welds made in inert atmospheres on light alloys.
- Tentative methods of testing weld metal deposited by submerged arc welding (mild steel).
- Particular instances of fatigue failures (5 documents).
- Note on the IIW enquiry on fatigue failures in service.
- Fabrication of welded sheet constructions by the rolling-up method.
- Nomenclature of the welding process.

In addition, two documents will shortly be issued in the form of publications of the IIW itself:

- Collection of reference radiographs of welds in aluminium and its alloys and in magnesium and its alloys.
- "Resistance Welding" section (in 17 languages) of the Multilingual Collection of Terms for Welding and Allied Processes.

Furthermore, eight radiographs have been added to the Collection of Reference Radiographs of Welds on Steel which is one of the publications of the IIW.

(*b*) *Relations with ISO and ILO*

A recommendation was made for the transmission to ISO/TC 11 SC 3 "Unification of boiler codes—Welded

construction" of a document giving information on the provisional values of image qualities necessary in radiography of pressure vessels in steel for thicknesses lower than 200 mm (8 in.)

In addition, the IIW appointed an observer at the meetings in Helsinki of the plenary committee of ISO/TC 44 "Welding".

With regard to the International Labour Office, arising out of previous contacts between it and the Executive Council of the IIW, Commission XIV "Welding Instruction" is to collaborate with ILO in the preparation of recommendations on the training of welders and welding instructors in accordance with a programme proposed by ILO; a new working group will deal with this question.

(c) *Terms of Reference of the Commissions*

Following upon the decisions taken last year concerning the study of certain new welding processes, arrangements have been made to clarify the division between Commissions II "Arc Welding" and XII "Special Arc Welding Processes" of the study of different processes and methods of arc welding. In particular, it was decided that, if metal-arc welding with continuous covered electrodes or with the mechanical feeding of covering is within the terms of reference of Commission II when there is no gas shielding, this process used with gas shielding should fall within the sphere of activity of Commission XII.

Furthermore, metal-arc welding with stationary covered electrode (or firecracker welding) and metal-arc welding with continuous bare electrode (without shielding), which were previously within the terms of reference of Commission XII, are now to be within those of Commission II.

(d) *Setting up of a Sub-commission and Working Groups*

A Sub-commission was set up within Commission XII "Special Arc Welding Processes" to study problems relating to new special arc welding processes. In addition, apart from working groups composed of members of Commissions, it was decided to set up an IIW group for the study of arc physics; this group is of interest both to Commission II "Arc Welding" and Commission XII "Special Arc Welding Processes". The preparations for its formation will be the responsibility of the Executive Council of the IIW and the two Secretaries who will get in touch with a certain number of the authors chosen to present papers at the meeting on the theme "Arc Physics" which the Institute of Welding in the United Kingdom is to hold in the Autumn of 1962.

(e) *Houdremont Lecture*

The Houdremont Lecture was this year delivered by Professor Cohen (United States).

Exhibitions of Publications and Documents

During the Assembly much interest was shown in a display of all the publications of the IIW and of copies of all the documents emanating from the Commissions which have been recommended for publication in the journals of member countries.

A Catalogue of new publications on welding and allied processes published throughout the world and exhibited for the first time at Opatija in 1959 and Liège in 1960 has been prepared by the Belgian Organising Committee for the Liège Assembly with the collaboration of the Yugoslav Organising Committee for the Opatija Assembly. This document, which was exhibited at New York and has been distributed to the national delegations, completes the first four sections of the Catalogue which, taken as a whole, constitutes a general bibliography on welding and the allied processes.

AWS Annual Meeting

No public session or lectures, apart from the Houdremont Lecture, were organised during the IIW Assembly because it overlapped with the Annual Meeting of the AWS which was held from 17th to 21st April. At the invitation of the AWS the majority of IIW participants attended various items on the programme of this meeting which included as well as a banquet on 17th April, plant tours and visits to the Welding Exposition, the presentation of numerous papers; nine of these were given by non-American members of the Commissions of the IIW and dealt with aspects of the work of the Commissions concerned.

The titles and authors of the papers concerned are as follows:

"Notes on the appreciation of brazing filler metals and fluxes" by C. G. Keel (Switzerland), G. M. A. Blanc (Switzerland), and J. Colbus (Germany)

"Gas mixtures in shielded-arc welding with consumable electrode" by M. M. Komers (Germany) and L. Wolff (Germany)

"The solutions adopted in some difficult applications of flash welding" by M. E. Bylin (Sweden)

"Evaluation of the fatigue limit of welded steel assemblies using a single welded specimen or test element" by R. Cazaud and H. de Leiris (France)

"The calculation of welded connections" by A. Van Douwen (Netherlands)

"Pressure vessel design requirements in the future" by W. B. Carlson (United Kingdom)

"The role of hydrogen in arc welding with covered electrodes" by N. Christensen (Norway)*

"Influence of residual stresses and metallurgical changes on low-stress brittle fractures in welded steel plates" by A. A. Wells (United Kingdom)*

"Inspection and measurements of properties of welds on plastics" by G. A. Homés (Belgium)

In addition the AWS chose, from among names which it asked the IIW to put forward, Mr. H. Granjon (France), to give the AWS Adams Lecture for 1961 and Professor N. N. Rykalin (USSR) to present two papers in the AWS Educational Lecture series.

The title of the Adams Lecture was "Studies on Cracking of and Transformation in Steels during Welding", while that of the Educational Lectures was "Heat Sources, Heat Flow and Heat Effects in Welding".

Visit to Canada

On 20th April many participants in the IIW Assembly took part in an expedition to works in the Niagara Falls area. This expedition was arranged by the Canadian Council of the IIW which supported the AWS in the organisation of the Assembly.

1962 Annual Assembly in Oslo

The 1962 Annual Assembly of the IIW will be held in Oslo, Norway, from 25th to 30th June. The programme will follow the usual pattern and will include a formal opening session, meetings of the Governing and Executive Councils at the beginning and end of the Assembly, the Houdremont Lecture, and four days devoted to meetings of the fifteen technical Commissions and their Drafting Committees. In addition, there will be a series of four lectures on "*High Pressure Pipelines and Penstocks*", to be held on 25th, 26th, 27th and 30th June. The lecturers will come from France, the United States and the U.S.S.R., and these sessions will be open to all participants.

An exhibition of publications from the various member countries of the IIW, featuring all kinds of welding literature and standards, is being arranged, and there will be an additional international exhibition of publications on the theme "*Design of Welded Constructions*".

During the week following the Assembly, from 31st June to 6th July, round trips will be organised, so as to enable participants to see Norwegian industry and research laboratories.

* Published in the April 1961 issue, volume 40, number 4, of *Welding Journal*.

Any further information may be obtained from the Honorary Secretary, Joint Committee for International Welding Relations, 54 Princes Gate, Exhibition Road, London, S.W.7.

Future Assemblies

The Governing Council of the IIW has accepted an invitation from the Finnish Welding Society to hold the 1963 Assembly in Helsinki.

EDUCATION

An increasing awareness of the need for instruction in welding design and practice is shown by the following notes on courses that have been arranged by several technical colleges in the London and Birmingham areas.

Those interested should obtain further details and application forms from the Principals of the Colleges concerned.

Welding Technology for Structural Steelwork

An evening course of ten special lectures on "*Modern Developments in Welding Technology and their Application to Structural Steelwork*" is being given in the Building and Civil Engineering Department of Twickenham Technical College on Tuesday evenings from 7.00 p.m. to 9.00 p.m. This course commenced on 17th October, and the lectures were by specialists from industry covering Welding Processes and Procedures, Welding Plant and Accessories, Types of Joint, Welding Metallurgy, Faults and Distortion, Methods of Testing, Specifications, Recent Developments.

The course is intended for engineers and designers of Graduate or H.N.C. status, and will be followed by a supplementary course commencing Tuesday, 16th January, 1962, on "*The Practical Design and Detailing of Modern Welded Structures*".

Further particulars may be obtained from the Twickenham Technical College, Egerton Rd., Twickenham, Middx.

Civil and Mechanical Engineering

A course on "*Steel Bridge Design*" (23 meetings) and one on "*Design of Rigid Jointed Multi-storey Steel Frames*" (11 meetings) commenced on 2nd October at the Northampton College of Advanced Technology, St. John Street, London, E.C.1.

A third course on "*Plastic Design in Steel*" (11 meetings) with Mr. E. Goodwin as lecturer, will commence at the College on 8th January 1962.

A similar course "*Plastic Design of Structures*" with lectures by Mr. J. Corbett, commenced at the Birmingham College of Advanced Technology, Gosta Green, Birmingham, on 2nd October (20 weekly meetings).

Welding Techniques

Five full-time meetings on "*Recent Developments in Welding Techniques*", with various lecturers, are to be held on 5th and 6th April 1962 at the Wolverhampton and Staffordshire College of Technology, Wulfruna Street, Wolverhampton.

Design Appreciation

Also of interest to engineers responsible for design will be an Engineering Executives' Course on "*Design Appreciation*" from 27th November to 1st December. This has been arranged by the Council of Industrial Design, 28 Haymarket, London, S.W.1, and is specially prepared for engineering management. It aims to show the developing importance of individual design aspects in engineering, and to describe how improvement in design can be made.

NEWS FROM INDUSTRY

Goliath Crane for Dungeness

A 40 ton goliath crane of welded box girder construction has been supplied to Whessoe Ltd. for use during the construction of the two reactor pressure vessels at the Dungeness Nuclear Power Station.

With a span of 200 feet and a 65 ft. height of lift, the crane is notable for the use of a triangular section welded box girder; this is cheaper to manufacture and weighs less than the corresponding conventional rectangular section.

The crane, which is illustrated on the front cover, has been manufactured by J. H. Carruthers & Co. Ltd., of East Kilbride, who have specialised since 1954 in the construction of cranes using welded box girder sections.

The Manchester Joint Research Council

The Manchester Joint Research Council was established in 1944 as a co-operative effort between the University of Manchester (which includes Manchester College of Science & Technology) and the Manchester Chamber of Commerce. The broad objective of the Council is to further the contacts between science and technology, on the one hand, and industry on the other.

The Council has organised a number of lectures, discussion meetings and conferences designed to spread ideas and understanding concerning some of the many aspects of the relations between science and industry. In addition it has established a liaison service between local industry and science and technology, and information on the scope of this service is included in the pamphlets recently issued by the Council under the titles "*Sources of Scientific and Technical Information available to Local Industry*" and "*An Account of some of the Scientific and Technological Facilities which Manchester University and Manchester College of Science & Technology can offer to Industry*".

As part of the work of the liaison service, a free Technical Information Service now operates with the aid of a grant from DSIR to help firms with technical and other problems, by putting them in touch with suitable sources of information. These sources also include the fifty-two Research Associations and the DSIR Research Stations. Contact is maintained with these bodies so that the scope of their work and interests is known and enquirers can be advised accordingly.

In addition to Council members, a number of firms have appointed senior members of their staffs as "Associate members". These Associate members may like to suggest profitable subjects for investigation and discussion and also to seek advice from the Council. Although these activities are not limited to Associate members, it is hoped by the Council that firms may feel that they are doing something to further the future prospects of industry in the Manchester area through the appointment of senior members of their staffs as Associate members. Some Associate members have already offered their services to the Council in an advisory capacity and any further offers of this type would be welcomed. A nominal fee of £1 1*s*. per annum is charged for Associate membership of the Council.

Applications for Associate membership, copies of the two pamphlets mentioned above and all enquiries (including those relating to technical problems) should be addressed to The Executive Officer, Manchester Joint Research Council, c/o Manchester Chamber of Commerce, Ship Canal House, King Street, Manchester, 2. (Tel: DEAnsgate 5574).

C.E.G.B. Research Laboratories

Mr. L. Rotherham, Board Member for Research, recently opened the new N.W. Region Laboratories of the Central Electricity Generating Board at East Didsbury, Manchester.

The laboratories will, in general, be responsible for research studies which will lead to improved efficiency and availability of plant within the N. W. Region, and it is expected that much of the Board's work on welding will be done there.

The head of the new department is Mr. J. Harvey, Regional Research and Development Officer.

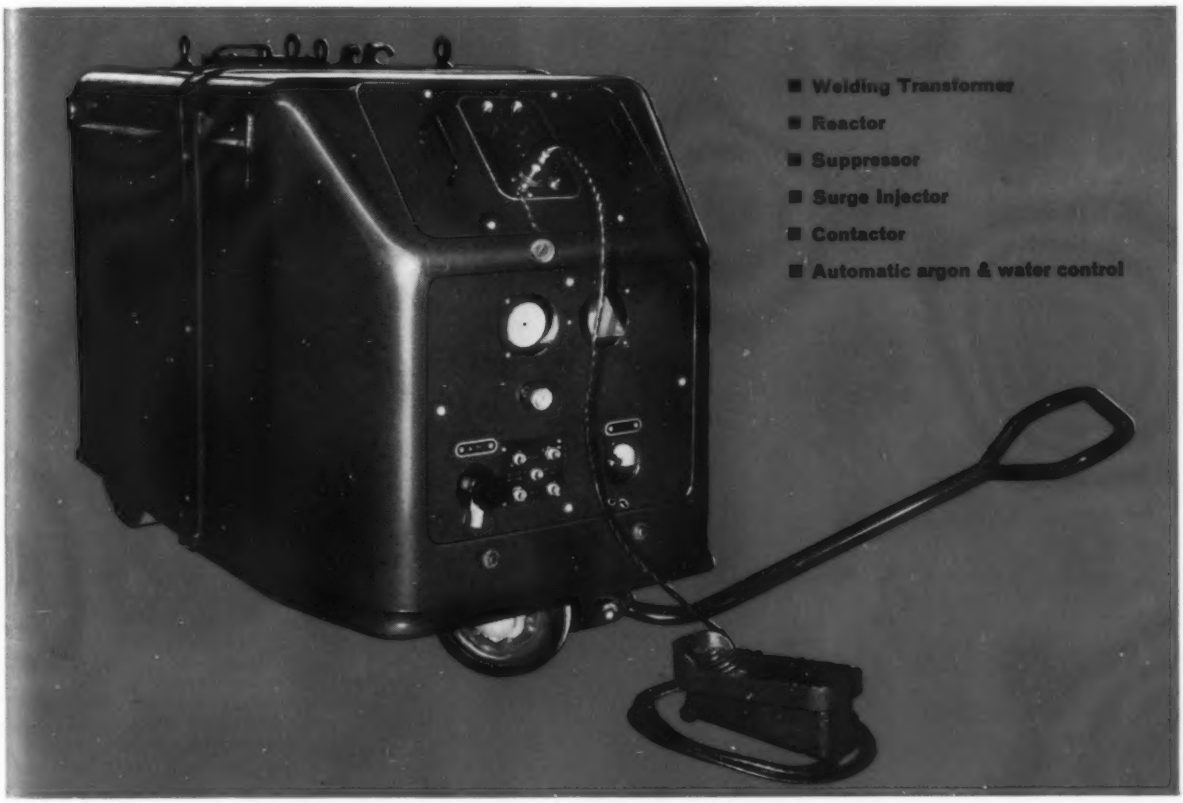
Avoiding distortion in cutting

Underwater cutting, for repair work on submerged structures, has several practical disadvantages, but has the distinct virtue of reducing thermal distortion to a minimum. From time to time this technique is adopted in normal fabrication, the practical difficulties

BRITISH OXYGEN

NEW COMPOSITE SET FOR ARGONARC WELDING

The British Oxygen new A.C. Composite Set type CPU. 300/80 incorporates all the necessary electrical gear for Argonarc welding in one case mounted on rubber tyred wheels for easy portability. It embodies many improvements, including a completely new and highly efficient surge injector and a power supply unit which delivers a high open circuit voltage of 80v. The CPU. 300/80 set also includes a D.C. Suppressor, contactor, automatic argon and water controls and ancillary electrical equipment. These improvements are rapidly establishing the new set as top of its class in aluminium and light alloy welding.



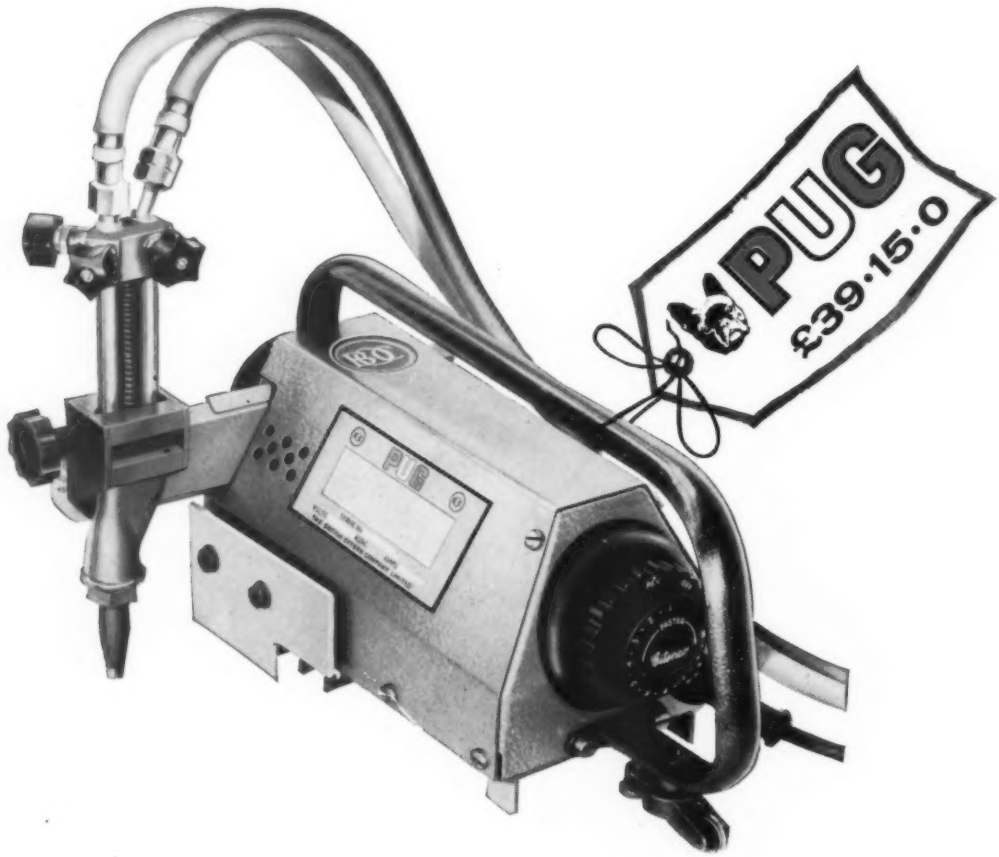
- *High open circuit voltage for excellent welding characteristics*
- *New surge injector for easy welding conditions*
- *Simple controls with easy-to-read dial*
- *Negligible television and radio interference*
- *Built-in power factor correction*
- *Suitable also for Manual Arc Welding*
- *Constructed to B.S.638 : 1954*

See the new A.C. Argonarc Composite Set at work. Write to arrange a demonstration and for full technical details.

THE BRITISH OXYGEN COMPANY LTD
Electric Welding Department, Quasi-Arc Works, Bilston, Staffs. Telephone: Bilston 41191

new PUG

low cost
light weight cutting machine



The light weight Pug is a new portable cutting machine made by British Oxygen. Tough and reliable, the Pug has important new features – making high quality mechanised oxygen cutting available to an ever-widening range of industries. The Pug is built for safety, reliability and long working life.

* *Cuts and bevels 2 in. steel.* * *3 in. to 45 in. diameter circles.* * *A.C./D.C. motor.*
* *Long trouble-free working life. Ask for demonstration.*

THE BRITISH OXYGEN COMPANY LTD

LIGHT INDUSTRIAL DEPARTMENT, SPENCER HOUSE, 27 ST. JAMES'S PLACE, LONDON, S.W.1

being reduced by the small 'depth' of submersion. An example has recently been brought to our attention by the British Oxygen Co. Ltd.

At the Consett Iron Co. Ltd. 1,500 convector plates $\frac{3}{16}$ in. thick were profiled with a 'Bison' cutting machine under $\frac{1}{4}$ in. of water. Although the cutting speed was thereby reduced by some 50% the technique proved to be economic by eliminating the need for straightening or flattening the plates after cutting.

European Directory

A four-language industrial information book 'ABC Europ Production' has been published by ABC Verlagshaus, Darmstadt, Germany, at a price of £3 8*s*. 0*d*. including postage. The directory includes extensive product indices to the names and addresses of some $\frac{1}{4}$ million firms. The main section contains about 10,000 article headings under which the individual manufacturers of seven European countries are recorded.

Welding Guide

Rockweld Ltd. have issued a concise reference guide to their range of electrodes, welding plant, and accessories.

Stud-welding Conference

A recent Overseas Licensees Reunion, organised by Crompton Parkinson (Stud Welding) Ltd. and their American associates, Gregory Industries Ltd., was attended by representatives from 15 countries.

The party first travelled to S. Wales to study manufacturing methods and to see an exhibition of stud-welding equipment and demonstrations of new techniques. This was followed by a two-day conference for the exchange of views on technical matters and a discussion of stud-welding applications.

Electric Stress Relieving Furnace

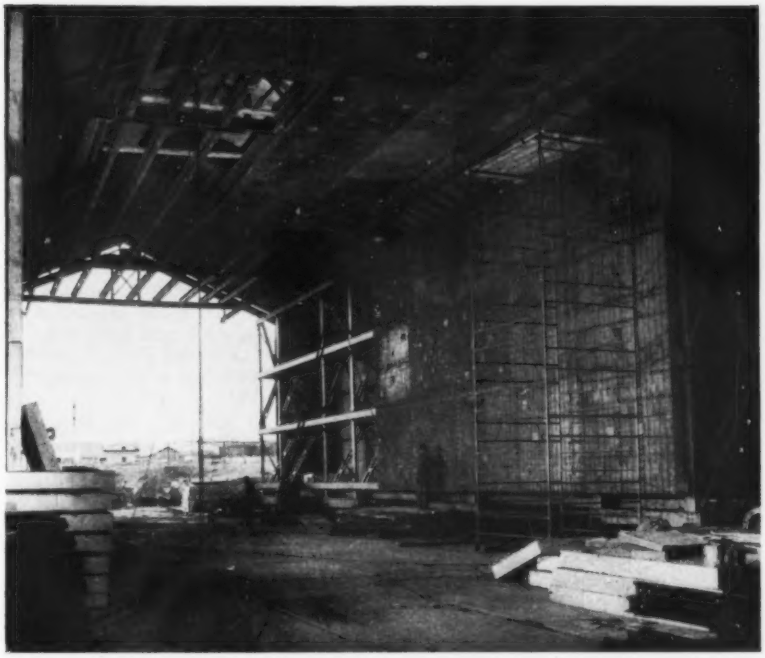
A large stress relieving furnace, 80 ft. long and 27 ft. square, for the treatment of nuclear heat exchangers and other large fabrications, has been erected at the Thornaby Works of Head Wrightson Teesdale Ltd.

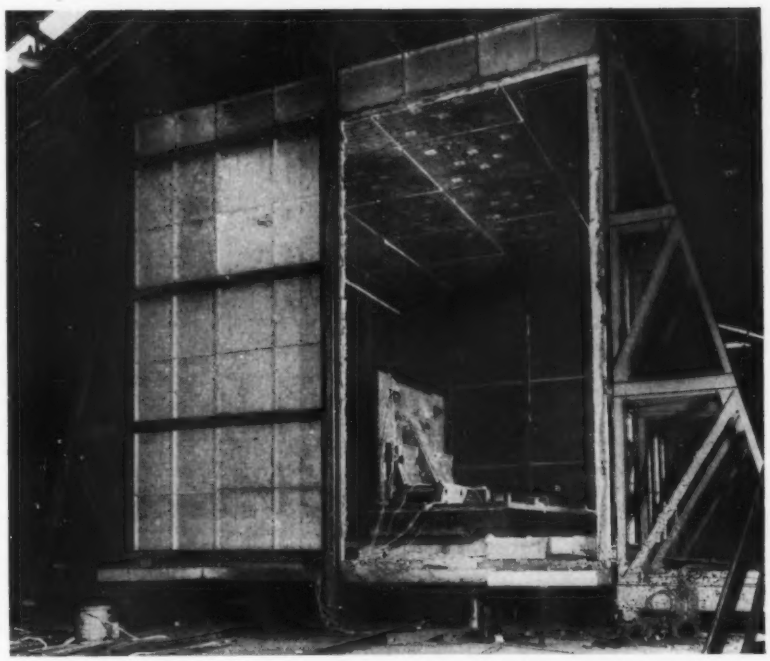
The furnace breaks away from traditional forms of construction; no firebrick is used, the hearth, walls and roof being constructed mainly of 8 ft. square mild steel panels filled with compressed mineral wool slabs.

The two illustrations show the general scheme of construction. The roof panels are fixed to joists under the roof truss; the walls are suspended from movable *A* frames running on transverse bogies; and the end walls can be lifted in two pieces by a mobile crane.

The workpiece is supported on cradles sitting on the furnace floor, which is also in panel form and is carried on bogies.

The furnace is of the closed type, so that scaling is reduced to a minimum.





Large stress relieving furnace at Head Wrightson Teesdale Ltd.

Heating is by a series of 40 kW units of 12 coiled elements in sintered alumina sleeves supported in stainless steel channels. The maximum consumption is 1500 kW at mains voltage. The design was by Cooper Electroheat Ltd., who also supplied the furnace panels and heating units. Head Wrightson constructed the bogies, cradles and supporting steelwork and erected the furnace.

Ferro Alloy Industry in India

The National Metallurgical Laboratory, Jamshedpur, is organising a Symposium on Ferro Alloy Industry in India, to be held early in February 1962. Details can be obtained from the Director of the Laboratory, Jamshedpur 7, India.

Piping systems for petroleum industry

After a long period of research and discussion the British Standards Institution has issued BS. 3351:1961 "Piping systems for the petroleum industry". The specification represents, in terms of British materials and manufacturing processes, standards of design and construction basically similar to those laid down in the American Standard Code for Pressure Piping (ASA B31.3, "Petroleum refinery piping").

Basic requirements and recommended practice are given for the selection and application of materials and components of piping systems in the petroleum and petrochemical industries; the design, fabrication, installation and testing of these systems are also covered. Among systems not covered by the standard are those in steam and power plants, for the design and construction of which British Standards or other accepted national codes are available.

Copies of the standard may be obtained from the British Standards Institution, Sales Branch, 2 Park Street, London W.1, price 30*s*. each. (Postage is charged extra to non-subscribers.)

Welding of steel pipelines

Lack of uniformity in welds employed for joints in pipelines and pipe assemblies carrying fluids and for securing flanges and attachments, such as anchor plates, to pipes can result from variations in practice as regards weld preparation, welding and heat-treatment procedures, the training and testing of welders and the inspection and testing of production welds. Good practice in all these respects having now been established, a series of British Standards is being issued to cover the production of welds in pipelines and pipe assemblies of any material used sufficiently widely to warrant its inclusion.

BS. 2971:1961 "Class II metal-arc welding of steel pipelines and pipe assemblies for carrying fluids" is the fourth standard in this series. It covers shop and site metal-arc welding of joints in carbon steel pipelines and pipe assemblies up to and including 24 in. dia. carrying fluids suitable for Class II conditions, and also the standards of workmanship of which the welders shall be capable and the standards to which the welds shall conform. It deals with butt welding, with or without a backing ring, where the welding is done from the outside only, and the welding of gusseted bends, sleeve joints, branches and flanges.

Copies of the Standard may be obtained from the British Standards Institution, Sales Branch, 2 Park Street, London W.1, price 12*s*. 6*d*. each. (Postage is charged extra to non-subscribers.)

STUB ENDS

▶The Aluminium Development Association has issued a new edition of its "Directory of Members with Index of Products". Copies can be obtained from the Association free of charge to bona fide enquirers.

▶Associated Electrical Industries Ltd. has recently issued a new publication "The AEI Arc Welding Guide", which replaces previous Metrovick Electrode Guides. Copies of publication 1891–71 can be obtained free on request from any AEI District Office.

▶The address of the Manchester office and warehouse of Norton Grinding Wheel Co. Ltd. is now Wharf Road, Sale, Cheshire (Tel. Sale 8285/7).

▶Among the exhibitors at the National Maintenance Exhibition, at Central Hall Westminster, on 13th–16th November, will be Eutectic Welding Alloys Co. Ltd., who will demonstrate maintenance and production welding equipment, and Metco Ltd., who will give working demonstrations of hardfacing.

▶Optoshield Ltd. have transferred their production and development departments to Watford. The Sales Dept. and Head Office remain at 146, Clerkenwell Road, London, E.C.1.

▶Dowty Rotol Ltd. have recently developed a new hydraulically powered vibrator for fatigue and vibration testing. Electrical signals are hydraulically amplified to operate the vibrator. This means that tape-recorded stress fluctuations in structures operating under service conditions can be reproduced for test purposes.

▶A recently published booklet "Booth Welding" illustrates 50 years of progress in the fabrication of welded steel tanks, pressure vessels and other products by John Booth & Sons (Bolton) Ltd.

▶The Visco Engineering Co. Ltd. has changed its name to VISCO Ltd.

▶The Welding Research Council of the Engineering Foundation, New York, has now moved to its new premises in the United Engineering Centre, 345 East 47th Street, New York 17, New York, U.S.A.

DIARY

30th Oct.–3rd Nov.—Symposium on Welding in Shipbuilding (Memorial Building, Institute of Marine Engineers)

1st Nov.—Institute of Welding—Annual Dinner (Connaught Rooms, St. Queen Street, 7.15 p.m.)
South London (Medway)—"*Terminal harbours and trans-continental pipeline*" by N. W. Grey (Fred. Braby & Co. Ltd., Crayford, 7.30 p.m.)
Manchester—"*Metallurgical aspects of the welding of heavy sections*" (Joint meeting with Met. Society) (Manchester Literary and Philosophical Society, George St., Manchester, 7.15 p.m.)

2nd Nov.—Institute of Welding—Presidential Address by H. West (Memorial Building, Institute of Marine Engineers, 6.0 p.m.)
N.E. Tyneside—"*Computer control flame cutting in shipbuilding*" by T. McIver.

6th Nov.—South Western—"*A practical approach to the problem of distortion*" by J. S. Waring (Radiant House, Bristol, 7.15 p.m.)

8th Nov.—N.E. Tees-side—Open discussion (Cleveland Scientific and Technical Institution, Middlesbrough, 6.15 p.m.)
South London—"*Fatigue and its effect on welding design*" by T. R. Gurney (54 Princess Gate, London, S.W.7, 7.30 p.m.)

13th Nov.—Sheffield—"*Transportation pipelines*" by R. J. Wright (Montgomery Hall, Surrey St., Sheffield, 7.15 p.m.)

13th–18th Nov.—International Factory Equipment Exhibition and Conference (Earl's Court)

14th Nov.—Leeds—"*The distortion problems in oxygen cutting*" by K. D. S. Semper (Hotel Metropole, 7.30 p.m.)

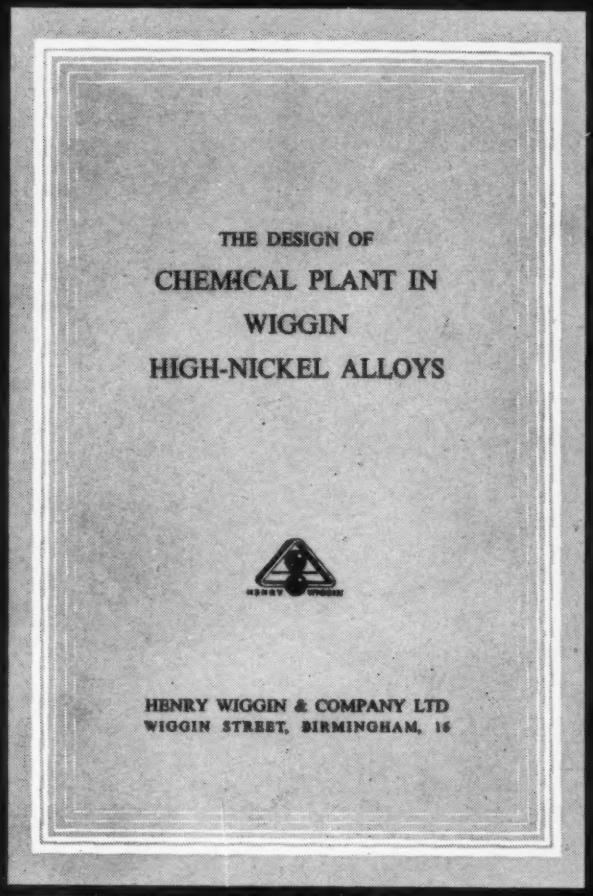
15th Nov.—Wolverhampton— "*Leaves from my notebook*" by R. G. Burt (Wulfrunians Club, 7.30 p.m.)
West of Scotland—"*Who teaches the science of welding?*" by Dr. J. Orr (Joint meeting with Inst. Structural Engineers) (Inst. Eng. and Ship. in Scotland, Glasgow, 7.0 p.m.)
East of Scotland—Film Evening (25 Charlotte Sq., Edinburgh)
Liverpool—"*Metallography of weld structures*" by Dr. S. J. Kennett (Central Libraries, William Brown St., Liverpool, 7.30 p.m.)

16th Nov.—Southern Counties—"*Welding in marine engineering*" by J. A. Dorrat (Portsmouth)

20th Nov.—Institution of Structural Engineers—"*Economics of framed structures*" by L. R. Creasy (New Civil Engineering Building, Liverpool University, Brownlow Hill, 6.30 p.m.)

22nd Nov.—North London—"*Quality and quantity applied to aero-engine welding*" by E. T. Hall (54 Princes Gate, London, S.W.7, 7.0 p.m.)

28th Nov.—Leeds—"*Chemical plants*" by F. S. Dickinson (Hotel Metropole, 7.30 p.m.)



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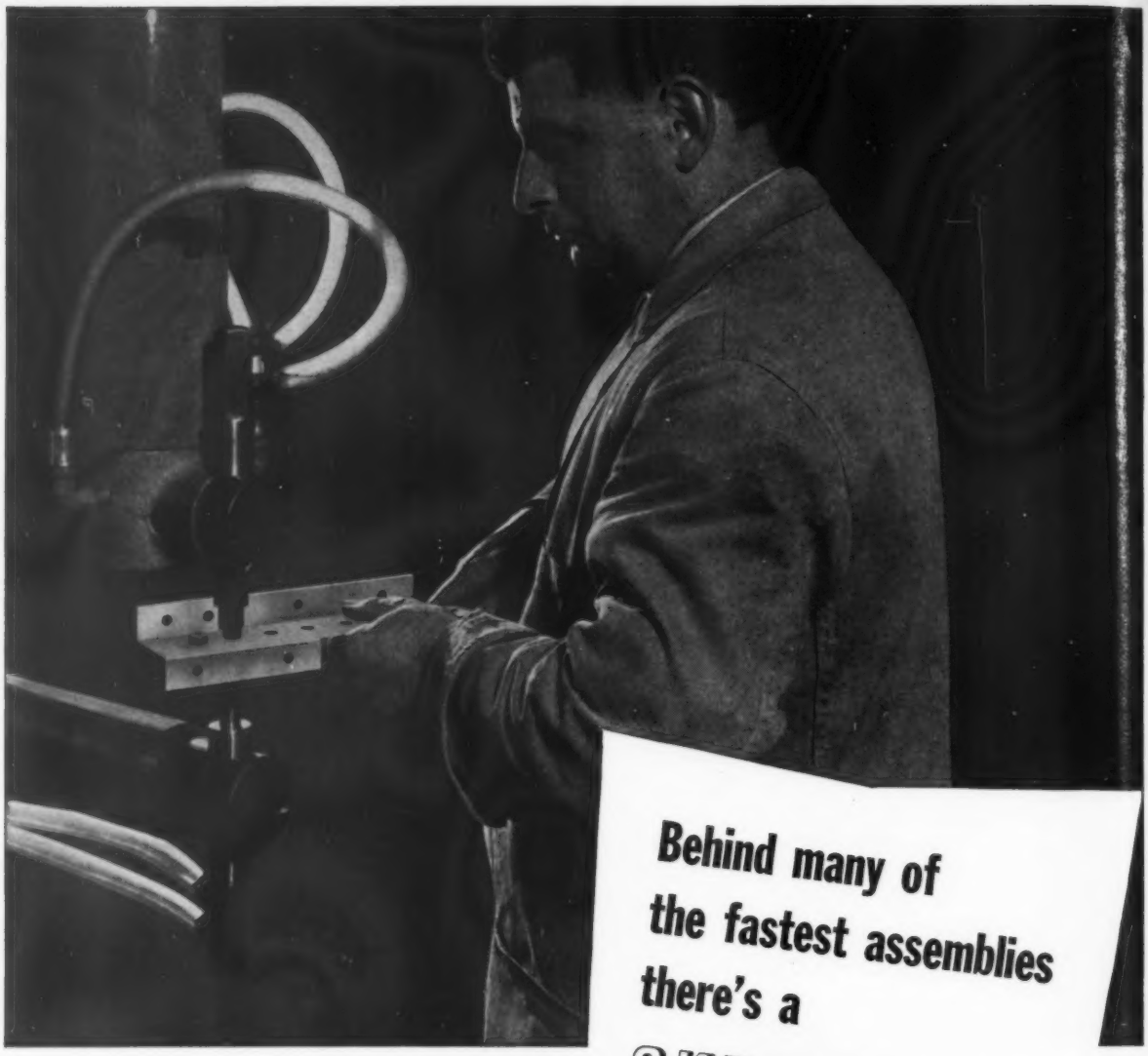

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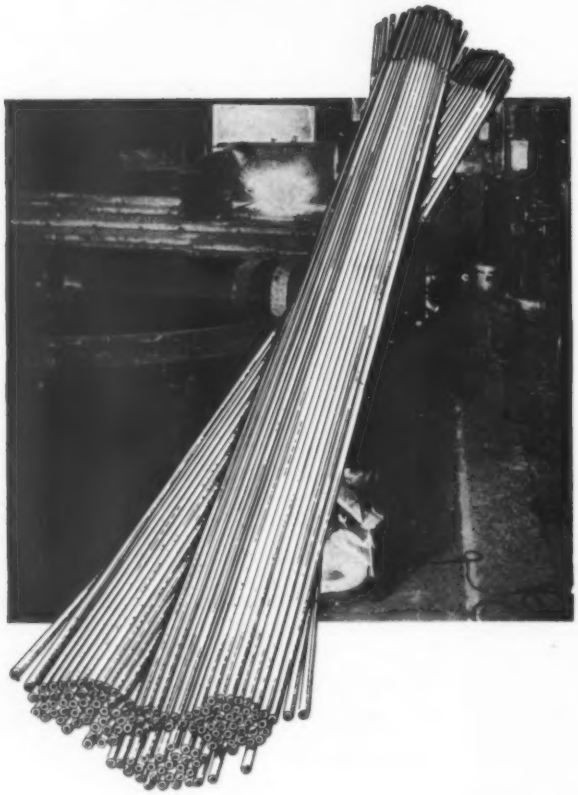

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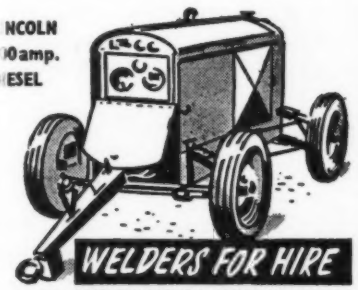

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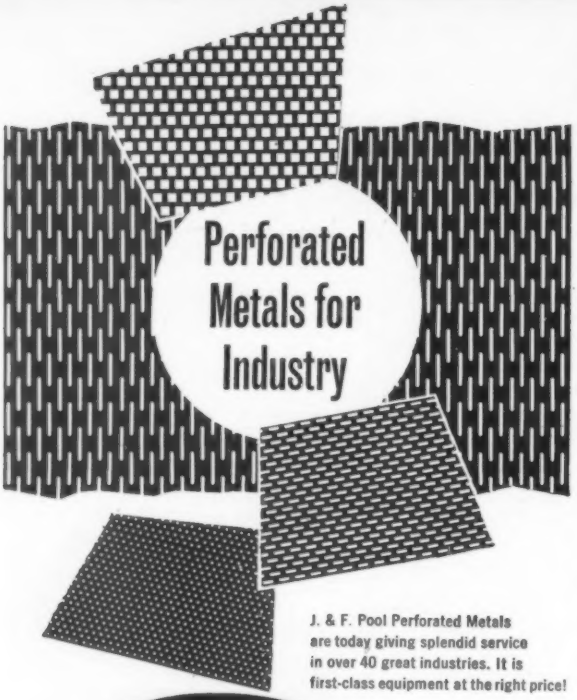

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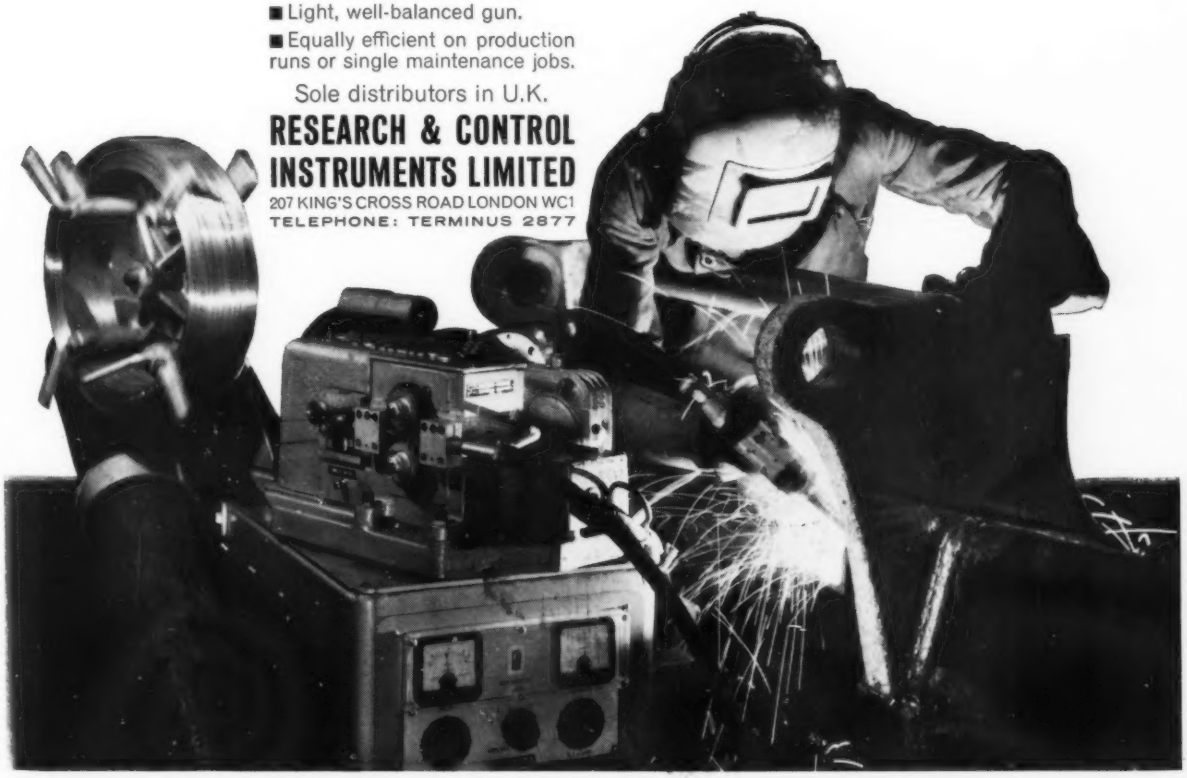

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